BRITISH WELDING JOURNAL

UNIVERSITY OF MICHIGAN
MAY 18 1959
ENGINEERING LIBRARY

1959 MAY

VOL. 6 NO. 5

CONTENTS

Oxygen-Cut Titanium
By J. C. Borland and W. G. Hull

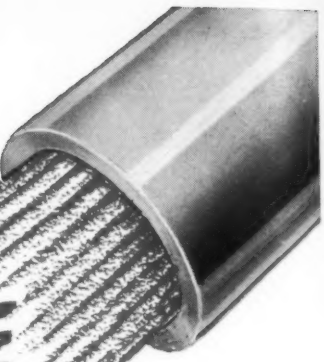
Welding and Lloyd's Register of Shipping
By G. M. Boyd

Welding and the Nuclear Power Programme
By Sir Leonard Owen

Effects of Hydrogen in Wrought Steel and in Ferrous Weld Metal
By P. D. Blake and W. I. Pumphrey

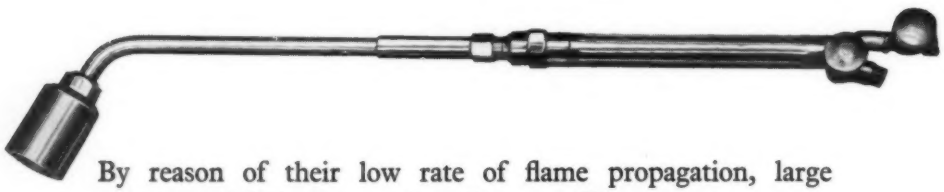
Phase Analysis of Resistance Welded Joints in Unalloyed and Low Carbon Steels
By D. Hrivnak and I. Hrivnak

the new B.I.G.



INTENSE HEAT

Propane Heating Torch



By reason of their low rate of flame propagation, large volumes of fuel gases such as Propane, Pyrogas, etc., can be safely mixed with Oxygen without the danger of the flame back-firing. The larger the volume of gases mixed, the greater the flame and the resulting heat value. The new B.I.G. PROPANE Heating Torch uses this principle with truly remarkable results.

Where large volumes of heat are required this portable unit with its terrific heat has unlimited applications in every Engineering Industry.

British Industrial Gases Limited

700, GT. CAMBRIDGE RD, ENFIELD, MIDDX. Telephone: **ENField 4022,** Telex: **24128**

Sales and Technical Assistance available in most areas

NRP 9023

Inside front cover

BRITISH WELDING JOURNAL

OFFICIAL JOURNAL OF THE INSTITUTE OF WELDING AND THE BRITISH WELDING RESEARCH ASSOCIATION

Incorporating the *TRANSACTIONS* of the Institute and *WELDING RESEARCH*

MAY 1959
Volume 6 Number 5

Subscriptions
Members free:
Extra copies at 7*s.* 6*d.* each
Non-Members:
£5 per annum
Single copies 10*s.* each

Published Monthly

Editor-in-Chief: G. PARSLOE, M.A.
54 Princes Gate, London, S.W.7
(Knightsbridge 8556)

Executive Editor:
C. ROWLAND HARMAN, B.Sc. (Eng.)
Editech
6 Ridge End, Hook Hill Lane
Woking, Surrey
(Woking 2981)

Advertisement Offices:
4 Grosvenor Gardens, London, S.W.1
(Sloane 0063)

Note: Editorial contents copyright. The Institute of Welding and the British Welding Research Association do not hold themselves responsible for statements made or opinions expressed in articles appearing in this Journal.

CONTENTS

INSTITUTE OF WELDING

2nd Annual Lecture

197 **Welding and the Nuclear Power Programme**
By Sir Leonard Owen

205 **Welding from the Viewpoint of Lloyd's Register of Shipping** *By G. M. Boyd*

211 **Effects of Hydrogen in Wrought Steel and in Ferrous Weld Metal** *By P. D. Blake and W. I. Pumphrey*

BRITISH WELDING RESEARCH ASSOCIATION

225 **Metallurgical Examination of Oxygen-Cut Titanium** *By J. C. Borland and W. G. Hull*

229 **Phase Analysis of Resistance Welded Joints in Unalloyed and Low-Alloy Carbon Steels**
By D. Hrivňák and I. Hrivňák

238 **NEWS AND ANNOUNCEMENTS**

244 **CURRENT WELDING LITERATURE**

The Institute of Welding 54 Princes Gate, London S.W.7

British Welding Research Association 29 Park Crescent, London W.1

Cut and come again

In high production belt grinding, costs are determined by the speed with which the job is done; that is, the cutting efficiency of the abrasive belt.

Behr-Manning RESINALL METALITE BELTS

. . . offer in full the well-known advantages of resin bonding, standing up to all the heat and strain you can give them. They combine with this a remarkable resistance to wear. Try the effect of the exclusive double bond of thermo-setting resin and see the extra life it adds to the consistently fast cutting action. More important, try the effect on your overall grinding costs — and *profits*. See your Norton representative or write direct to us.

BEHR-MANNING Coated Abrasives

Made by Behr-Manning Ltd., Belfast, and marketed in the U.K. by their associated company

NORTON GRINDING WHEEL COMPANY LTD.

COATED ABRASIVES DIVISION, WELWYN GARDEN CITY, HERTS. Tel: Welwyn Garden 4501 (10 lines)

NORTON *and* BEHR-MANNING *factories also in Argentina, Australia, Brazil, Canada, France, Germany, Italy, Northern Ireland, South Africa and U.S.A.*

NORTON* ABRASIVES

* *Trade Mark of Norton Grinding Wheel Co. Ltd.*

NGW CM 150

ANNUAL LECTURE 1959

Welding and the Nuclear Power Programme

By Sir Leonard Owen, C.B.E., M.Sc. M.I.C.E. M.I. Mech. E., M.I.Ch.E.

WHEN the United Kingdom embarked on a programme of nuclear power to form a substantial proportion of the British power installations it was recognized that large quantities of radioactive materials would be produced. In presenting the programme of nuclear power the Atomic Energy Authority were also expressing their confidence that these radioactive materials could be contained without hazard to anyone working in the plants or living near to them. From the production of this radioactivity, and the public duty to contain it, arise differences between plant for conventional usage on the one hand, and for nuclear power on the other. In conventional practice facilities are rarely denied for inspection and maintenance during the operational life of a plant, but in nuclear work this frequently happens because of radioactivity. Designing for this sort of life clearly introduces requirements for which no Code of Practice caters at present. All techniques to be used must be subjected to the closest scrutiny so that safety can be assured over long periods of usage. Well tried and trusted materials may be called upon to withstand more rigorous conditions than ever before, or the development of new materials may be necessary to enable them to be used for the first time as engineering materials.

Frequently, materials with superlative qualities in some direction, such as high-temperature strength or corrosion resistance, result from intensive development but prove to have no engineering advantage because there is not available at the same time a practical joining process which can be applied without causing deterioration in the joint or in the neighbouring parent material. For this reason the development of welding processes has received particular attention over the wide field of metals both ferrous and non-ferrous with which the nuclear power programme has been concerned. Although such a programme touches activities which may escape notice when the casual observer thinks only of power production, this account will be limited to problems of power reactors of present and future types. This does not minimize the importance of advances such as the positional welding of aluminium or stainless steel tubes for chemical processing plant, which have resulted from work in other fields.

THE CALDER HALL TYPE OF REACTOR

Fuel Elements

The principal welding problems of the fuel element in the early stages of research and development have been described previously, and the need to keep joining clearly in mind during the evolution of alloys of excellent corrosion resistance has been demonstrated.

General requirements

Because above 300°C. uranium reacts rapidly with carbon dioxide the fuel in a Calder Hall type of reactor must be enclosed in a gas-tight container or can, which will also contain fission products. Not only does this serve to prevent reaction between the gas and the fuel but it is designed to promote the efficient transfer to the coolant of the heat produced by fission. Inefficient heat transfer would result in a rise of temperature in the centre of the fuel element and this could lead to distortion from the phase transformation of uranium.

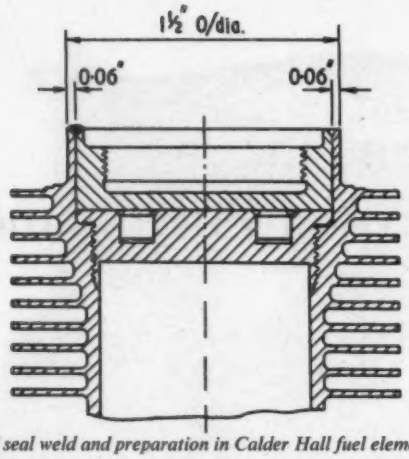
The can is therefore expected to remain gas tight during the life of the fuel element even though the fuel may, as a result of irradiation, increase in size and change its shape owing to growth, surface wrinkling, and swelling. In the first power producing reactors it must be able to accommodate strains which may, for example, result from an expected increase in length of 3% at temperatures below 300°C., or from an expected overall increase in size of 15% arising from gaseous fission product swelling at temperatures of about 450°C. after 3000 MWD/te.

The thermal expansion coefficient of magnesium is approximately twice that of uranium, so that separation of the can and bar will occur owing to differential expansion during loading, or subsequently during thermal cycling. Relative movement of bar and can is prevented by pressing, after welding, of the can body into grooves in the fuel rod. During this process, and

---

The second of the Annual Lectures was delivered at a special meeting of the Institute, held at 54 Princes Gate on Thursday, 12th March 1959.

Sir Leonard Owen is Managing Director of the U.K.A.E.A.
In the unavoidable absence of Sir Leonard Owen, the Lecture was read by Dr. A. B. McIntosh, Head of the Culchech Laboratories of the U.K.A.E.A.



1—*End seal weld and preparation in Calder Hall fuel element*

subsequently during operation, the end welds are subjected to stresses because the relative movement is prevented. During the life of the cartridge other stresses arise from the growth and swelling of the uranium fuel.

An acceptable design of container must not only have high efficiency of heat transfer to restrict the fuel operating temperature but the heat-transfer surface should be maintained over the full length of the can, otherwise the neutron flux peak near the end of the bars would cause a temperature peak. With stacked fuel elements this could lead to excessive creep deformation of the magnesium alloy end fittings.

Weld design and process

The need to carry the heat-transfer fins through to the ends of the cartridge has favoured the choice of an end edge weld, as shown in Fig. 1. When aluminium was used for the Windscale fuel element can the seal weld could be covered by braze metal to provide a double seal, but magnesium alloy cans are not easily brazed and reliance has to be placed solely on the weld. To meet the stresses involved the strength is increased by designing for a minimum penetration of 0·030 in. By shaping the inner edge of the can wall and the outer edge of the end cap fitting to provide a 45° edge preparation, this penetration is achieved. The tungsten argon-arc process is used, a machine having been developed for the purpose. The first of two passes is made with a short arc length (0·004 in.) to achieve adequate penetration, and the second with a long arc (0·05 in.) to obtain a satisfactory weld finish. The first run produces an irregular weld, but the short arc is necessary for adequate penetration. The second run removes the irregularities and produces a smooth weld, as shown in Fig. 2. The average throat thickness of the final weld is not less than 0·040 in. and is unlikely to fall below 0·030 in. at any point. Crater elimination at the end of welding is achieved by a controlled reduction in the welding current.

Further development work has produced a new welding machine which achieves greater control of all the welding variables. Welding current is maintained constant, regardless of variations in the mains voltage, and better arc stability is obtained by using surge injection in place of a superimposed h.f. voltage. Using an impervious shroud a controlled addition of 0·1 to 0·15% nitrogen to the high purity argon gives improved arc stability and a smoother weld without in any way impairing its properties.

Operational requirements

A Calder reactor charge consists of 10,000 fuel elements, and if in any cartridge the carbon dioxide should gain access to the uranium this must be discharged to avoid possible contamination of the reactor core by fissile material. The reactor would be closed down for several days, and several failures in a week would effectively close it down altogether. Even infrequent failures would be intolerable because of loss of plutonium output.

Inspection tests

The incidence of defective cartridges must be kept well below 0·1% a year and to guarantee this, 100% inspection with a sufficiently sensitive and non-destructive technique is essential.

To do this the cartridges are filled with helium, and subsequently tested for helium leakage with a mass spectrometer, with which a leak rate of the order of 0·01 lusecs* can be detected (Fig. 3). It has been shown experimentally that a leak rate of 1 lusec of helium under these conditions corresponds to a cylindrical hole 0·001 in. dia. and 0·03 in. long. Extremely fine pin holes are thus detected. Every cartridge is also radiographed.

Welds of the type described have also been extensively tested, and a typical value of the pull-out strength at 400°C. is 200 lb.

Performance

Experience to date has been sufficient with such fuel elements to show that the welds have given satisfaction, for the records show that, out of a total of 60,000 fuel elements used in three reactors, in only

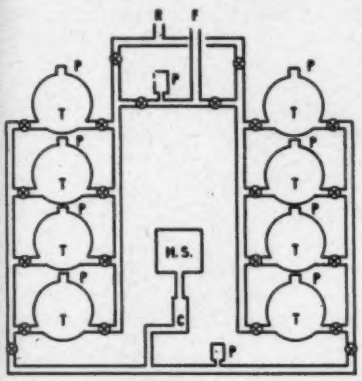
* 1 lusec equals a pressure rise of 1μ/sec per litre

2—*Edge weld on fuel element after second run*

Table I

Dimensions of commercial nuclear power stations in comparison with Calder Hall

	U.K.A.E.A.	*A.E.I.–John Thompson*	*G.E.C.–Simon Carves*	*Nuclear Power Plant Co.*	*Atomic Power Projects*
Location	Calder Hall, Cumberland	Berkeley, Gloucestershire	Hunterston, Ayrshire	Bradwell, Essex	Hinckley Point, Somerset
Shape	Cylinder	Cylinder	Sphere	Sphere	Sphere
Thickness	2 in.	3 in.	2¾ in.	3 in.	3 in.
Dimensions	37 ft dia. × 70 ft high	50 ft dia. × 80 ft high	70 ft dia.	66 ft 9 in. dia.	67 ft dia.
Net Output per Station	65 MW	275 MW	300 MW	300 MW	500 MW



MS. Mass Spectrometer
C. Cold trap
⊗. Valves
P. Pirani gauges
T. Test canisters
R. Roughing line
F. High vacuum connected to diffusion pump inlet *via* cold trap

3—*Layout of rig for production testing of completed cartridges by the mass spectrometer. (Each canister has an air bleed for ease in removing cartridge)*

seven of them have the welds been suspect during a full irradiation life. Of these only three could be attributed to weld failure arising from strains produced in the pile or from handling. The remaining four were shown by post-irradiation examination to be caused by leakage of gas through the end welds.

The Pressure Vessel

So much has been written about the construction of Calder Hall that no more need here be said than that it represented, at the time, the maximum extrapolation of existing practice considered acceptable, and gave a new datum line of experience from which the civil power stations could be conceived and designed. The natural development with the same basic materials was to larger pressure vessels, and Table I shows the increase in dimensions of the present Central Electricity Generating Board reactors compared with Calder Hall.

The site welding of such large vessels in thick plate is something new, but the pressure vessel manufacturers of this country have taken the problem in their stride, even though no existing Code of Practice covers the conditions. Anything more than a remote inspection of the pressure vessel during its operational life is unlikely, so that all welding and inspection practices must be examined to ensure unqualified confidence in the safety of the vessel. The vessel may be subject to a type of failure—brittle fracture—which is not yet sufficiently well understood to be allowed for simply in design.

Brittle fracture

Although some would claim that brittle fracture is an overrated consideration prudence has dictated that we base our views on a philosophy aimed at the least possible risk.

Some facts are indisputable. Firstly, there is a transition temperature below which a steel structure can rupture in a brittle manner and above which the fracture is ductile. Secondly, a ductile crack consumes a great deal of energy and usually, by comparison, travels slowly; and thirdly, a brittle crack requires relatively little energy and travels rapidly.

Three things are necessary to cause brittle fracture in a steel structure—a notch, a stress, and a temperature lower than the transition temperature. The notch, which may be a defect in the steel, a welding defect, or a sharp corner provides the stress concentrator to start the crack. The overall stress provides the energy to make the crack grow, and the lower the temperature the smaller need the crack be to propagate for any level of stress.

Thus at temperatures below the transition temperature a steel structure could operate without failure if there were no stress concentrators or if there were insufficient energy available to propagate such small defects as might be present. It is well known that many engineering parts, such as turbine rotors, have functioned successfully for years at temperatures where they are notch brittle, particularly if stress relieving has been carried out to relax internal stresses which may be at a high level. Safety of this type is conditional and depends not only on clean steel and good welding practice, but on the excellence of inspection.

Inspection of welds in thick plate

The inspection method must be sensitive enough to reveal and locate objectionable flaws so that they can be eliminated.

Freedom from gross lamination of the plate, particularly in the region of edges of cut-outs where welding is to be done, is easily established by an ultrasonic test or by magnetic crack detection, and such methods are not invalidated by increase in plate thickness. The drive to thicker plate, however, must always be balanced against the ability to inspect all welds, to repair following the inspection, and then to inspect the repair to prove that the weld is now as specified. Nothing less than this can ever be acceptable in a reactor pressure vessel.

X-radiography

Butt welds in plate up to 3 in. thick, the normal thickness in most of the reactors at present under construction, are inspected by radiography according

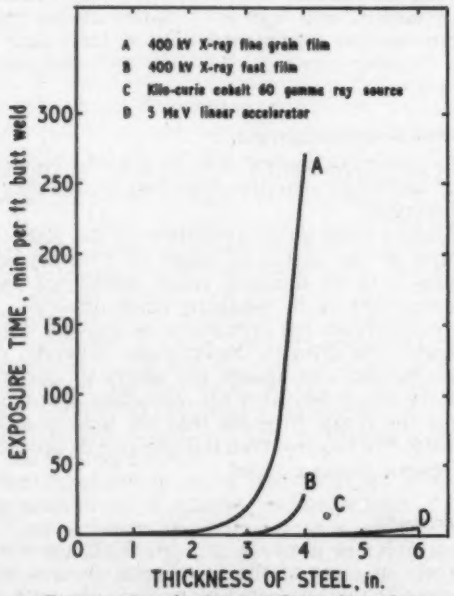
to some such code as BS.1500 for fusion welded pressure vessels, which requires a radiographic technique capable of showing a difference in steel thickness of less than 2%, *i.e.*, 0·060 in. for a plate of 3 in. thickness.

Radiographs of this quality are obtainable during site construction of pressure vessels made of 3 in. plate by using 400 kV transportable X-ray equipment. To obtain good resolution of defects it is desirable to use a fine grain X-ray film and to operate with the radiation source at a distance not less than 3 ft from the plate. The exposure time per 1 ft run of butt weld is about 15–20 min under these conditions.

Difficulties are encountered in applying the same type of radiographic equipment to butt welds in plate thicker than 3 in. because exposure times using the same techniques rise steeply, as shown in Fig. 4. To retain practical exposure conditions with this type of equipment, modifications in technique are necessary. Curve *B* in Fig. 4 shows the reduction in exposure time that could be achieved by the use of films of higher speed and coarser grain. To shorten the exposure time still further, X-ray operators may be tempted to reduce the distance between the source and the plate. Modifications of technique of this sort, aimed at regaining a practical exposure, will impair the standard of inspection. Even if long exposure times could be tolerated the increased proportion of the scattered to the direct radiation would reduce the standard of inspection, unless more penetrating higher-energy X- or gamma-rays were to be used. For site inspection in the immediate future multi-curie cobalt-60 gamma-ray sources of high specific activity provide the best prospects for maintaining the standard of site inspection.

Gamma radiography and linear accelerators

A few sources of kilo-curie strength and specific activities of 30–40 curies/gram are now in the possession of some of the civil reactor consortia. Exposure times per foot run of butt weld in 4 in. plate using fine grained film may be about 15–20 min, as shown by the point *C* in Fig. 4. In practise exposure times are likely to be 1–2 hr because these sources will have to operate at greater distances from the plate to maintain the standard of resolution of the defects, but to compensate for this, several feet of weld will be radiographed during each exposure.



4—Exposure times for various radiographic techniques

Alternatives to cobalt-60 sources are the megavolt X-ray generators. Of these, because of its high output, the 5 MeV linear accelerator is attractive. These units are not at present available in a form suitable for site radiography, but the Research and Development Branch of the U.K.A.E.A., Industrial Group, are piloting, in association with Central Electricity Generating Board, the Armament Research and Development Establishment, and the industrial consortia, the development of a linear accelerator for site use. The present indications are that such equipment could extend radiographic inspection of steel up to and possibly beyond 6 in. with the short exposure times indicated by curve *D* in Fig. 4.

Automatic positional welding

A typical gas cooled reactor vessel at present under construction will contain more than twenty tons of weld metal. Automatic welding is confined at present to sub-assemblies of plates which can be welded in the downhand position, amounting to about 15% of the welding. The best control of welding is obtained when inspection keeps pace with welding, but rarely does it prove an easy matter to programme the work to keep them in phase. For thick plate inspection, which involves the use of multi-curie gamma sources that restrict the movement of personnel in the area, there is a clear incentive for the development of automatic positional welding machines suitable for site use, and for the prefabrication of the largest possible assemblies.

Other inspection methods

Although success is expected from the development work, the minimum size of defect detectable will probably increase with the plate thickness, and radiographic inspection thus will become less critical for the thicker plates.

For this reason, and also because of the need for providing extensive shielding to avoid radiation hazards in the use of high-intensity radiation, other inspection methods are being developed. Of these the use of ultrasonics is the most promising. The sensitivity of ultrasonic inspection is not normally limited by any steel thickness that might be used in any future pressure vessel design. It is already a valuable supplementary method of inspection by locating defects and its application to the examination of welds in thick plate will increase if techniques to discriminate with certainty between the various types of defects can be developed. To avoid operator fatigue, automatic scanning is essential, and automatic recording of results is desirable.

With joints of more complicated geometry, such as points of entry of stand pipes and ducts, the problem is more difficult even though ingenious engineering design may give some relief by introducing butt welding forgings. Where this is not possible there remains the need for an inspection method from which will be derived the confidence commensurate with that

for the rest of the pressure circuit. A defective nozzle weld could lead to loss of coolant or to failure of the main shell from the propagation of a micro-crack in the weld deposit. Suitable inspection should establish freedom from communicating paths from inside to outside and from micro-cracks or other defects acting as objectionable stress raisers. There is an urgent need to develop sensitive leak tests using techniques similar to those developed by the Industrial Group for the inspection of large scale processing plant, where freedom from leakage to vacuum standards is essential.

The pressure vessel under stress and irradiation

Whilst careful design coupled with stress relieving may reduce stress concentrations, and meticulous inspection may ensure relative and limited freedom from flaws, too little is known about fracture initiation for any certainty to be felt that simply because a crack or defect is below a minimum size it cannot precipitate fracture. In practice there are always stress concentrations, welding flaws, and other faults whose part in initiating fracture is not to be doubted, and good practice serves only to reduce the occurrence and extent of these to a minimum. Even so, as the number of feet of welding increases there is a proportionate increase in the number of flaws and in the probability that one of them will be so disposed as to provide the initiation of a crack. The safety of the vessel below its transition temperature depends on the value of this probability.

An important factor which may be left out of account is the amount of elastic energy stored in the vessel under stress. In Fig. 5, which relates to spheres constructed in 3 in. plate stressed to 4 tons/sq.in., the energy required to produce an equatorial brittle crack is compared with the stored elastic energy, and with the energy needed for a ductile crack. It is shown that there is stored in the vessel more energy than is necessary to cause the propagation of an equatorial brittle crack.

For safety to be guaranteed, the vessel under stress must always be above its transition temperature, where the energy available is insufficient to propagate a ductile crack. There may be some uncertainty at present in deciding what the value of this temperature is, in view of the combined effect on the initial value of the transition temperature of such influences as variation in plate quality, welding, increase in thickness of plate, and irradiation damage. Provisional estimates to guarantee safety under these conditions may give a value of transition temperature which later proves to be excessive, but a precise specification must await the accumulation of sufficient experimental data.

In each reactor, typical specimens are placed at positions where conditions of temperature and flux are similar to those of the pressure vessel. Groups of specimens are periodically removed and give a direct indication of the state of the vessel at that time. Among the important features about which information is being gathered are the nature of the neutron flux and flux spectrum at various parts of the vessel, particularly where neutron beams emerge from channels, the importance of various parts of the spectrum on the mechanical properties of the steel, the effect at different temperatures and times of irradiation and the comparison of changes of transition measured on small samples with those measured on large samples. In each reactor, monitoring programmes will show whether or not such important effects due to thickness and irradiation are additive.

It is known that the transition temperature of a steel is raised more or less progressively, perhaps to a saturation value, as a result of irradiation. This must be taken into account in deciding what the transition temperature of a pressure vessel is likely to be at any time during its operational life.

THE ADVANCED GAS-COOLED REACTOR

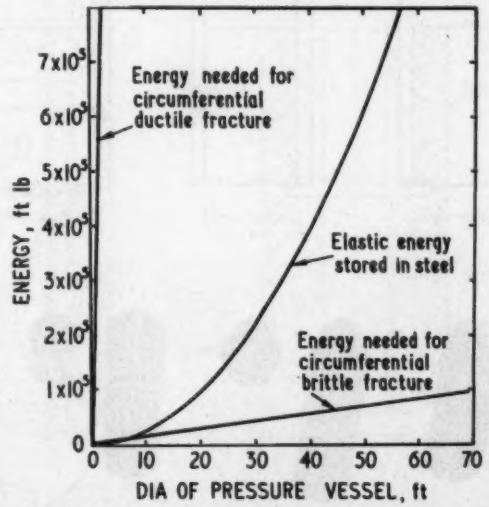
The largest fraction of the cost of electricity from nuclear stations arises from the capital cost of the reactor. On theoretical grounds the capital charge per unit sent out can be reduced by combining an increase in size of the reactor with increases in gas temperature and pressure.

Technically there may yet be advantages in increasing the size of a gas-cooled reactor still further but there is general agreement among nuclear power engineers that the practical limit has been reached and 'bloated' designs are viewed with disfavour.

Temperatures in the current C.E.G.B. reactors are limited because the natural uranium fuel should not reach the alpha–beta phase transformation, and because the magnesium is being used almost to its limit. Therefore in this type of reactor any increase in temperature is bound to be small.

To break through the limitations, the Atomic Energy Authority is building an experimental gas-cooled reactor of advanced design and higher rating—the A.G.R. project. This reactor uses new type fuel elements in which the uranium fuel is in ceramic form UO_2, whose melting point is about 2,800°C., and the magnesium is replaced by beryllium—a metal with low neutron capture cross section whose melting point is 1,280°C.

The beryllium canned fuel elements will enable the top gas temperature to be raised from about 400° to



5—Stored elastic and fracture energies of a spherical pressure vessel (3 in. thick plate stressed to 4 tons/sq.in.)

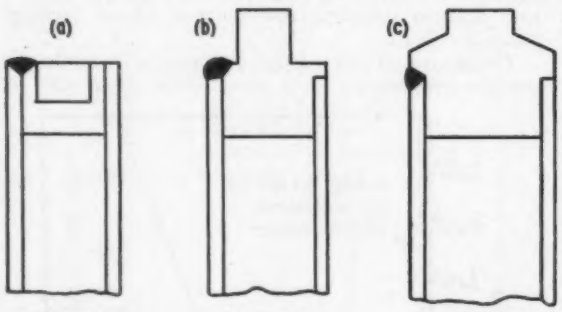
about 550°C., and among other things this allows the use of a reheat steam cycle raising the theoretical cycle efficiency by 5 or 6%.

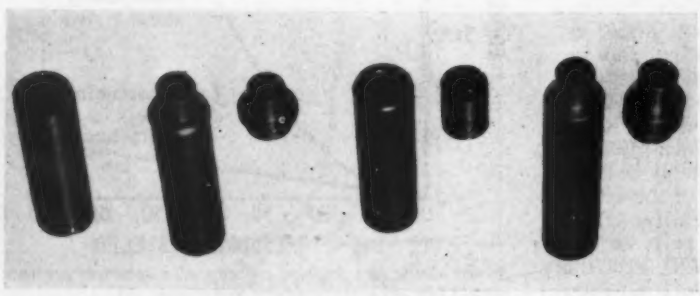
Welding of beryllium

The use of beryllium introduces a formidable set of processing difficulties which arise from its metallurgical structure and from its high reactivity towards oxygen and nitrogen. The finely divided metal and its oxide constitute an extremely grave health hazard, and the most elaborate precautions have to be taken to guarantee the protection of operators. The maximum content of beryllium metal dust or fine particles of oxide in the atmosphere must not exceed 2 millionths of a gramme per cubic metre, so that one ounce of dust could bring a volume of 500 million cubic feet or approximately 1 mile square 6 yards deep to the danger level of contamination. Complete enclosure of all equipment designed to heat the metal either in a vacuum or in an inert atmosphere is a first essential. Because the metal combines with the smallest traces of oxygen, the purity of the ambient atmosphere must be of a high order, say 99·995% argon. Normally the metal is fabricated by powder metallurgy methods, and techniques are now available for forming the metal accurately to the shapes required. So far attention has been concentrated on pressure and fusion welding.

Pressure welding demands absolute cleanliness of all the mating parts and complete control of the ambient atmosphere, together with a suitable choice of temperature and pressure variables to give the required reduction of cross sectional areas for the specimens to be joined.

Fusion welding again requires a completely inert atmosphere. The melting point of beryllium is not high and early difficulties were experienced owing to the development of cracks, associated with the poor ductility of the metal just below the melting point, coupled with the restraint imposed by joint design and cooling rate. Recent tests show that detailed attention to both these factors is an essential feature of processes for the production of crack-free welds. Both methods are being perfected and the final selection will, among other things, depend on the mechanical properties of the joint, particularly the creep strength and ductility in a carbon dioxide atmosphere and on the effect of irradiation on the weld metal. Figure 6 shows typical welds which are being made in beryllium.





6—Typical welds in beryllium and finished fuel cans

The pressure vessel

The increase in gas temperature has affected the design of the reactor pressure vessel. The principal change is to alter the flow path of the gas so that the pressure circuit—reactor vessel, heat exchangers, and inter-connecting ducting—is everywhere swept by the returning cool gas after its passage through the heat exchangers (Fig. 7).

Notable advantages follow. Firstly, the temperature of the returning gas will be about 250°–325°C., so that the whole pressure shell will be at a uniform temperature much greater than the maximum value to which the transition temperature of the steel can rise as a result of irradiation. The pressure vessel when under stress is consequently never likely to cool below its transition temperature and mild steel can thus be used. Secondly, the high temperature portions of the plant are limited in extent and relatively little stressed. The problem of possible creep does not arise.

By the incorporation of thermal shields in the reactor the neutron flux at the pressure shell is reduced to a value comparable with that in Calder Hall although in the core it is greatly increased. An internal neutron shield situated at the top of the core gives protection to the more complicated dome and to the welded support course with its duct openings. Inspection and maintenance are thus possible during the reactor's life.

In future nuclear power stations based on A.G.R. technology, the problems of reactor pressure shells may be eased as a result of these changes. A much greater power output is taken from a unit of reactor core volume so that the dimensions of the reactor vessels for a given power output will become smaller than for the existing designs of C.E.G.B. stations.

The technique of pressure vessel design and construction will not remain static. There is still great scope for improvement in almost every aspect of manufacture and construction, to produce a more perfect vessel at less cost. The ability to fabricate on site vessels with plate of increased thickness or of greater strength can always be put to advantage by the reactor designer. If increased vessel diameter is not required, gas pressures can be raised, so that for a given pumping power more heat can be transferred from the reactor to the heat exchanger. Alternatively, the pumping power itself, which detracts

from the overall efficiency, can be lowered for a given reactor heat-rating.

THE FAST FISSION BREEDER REACTOR

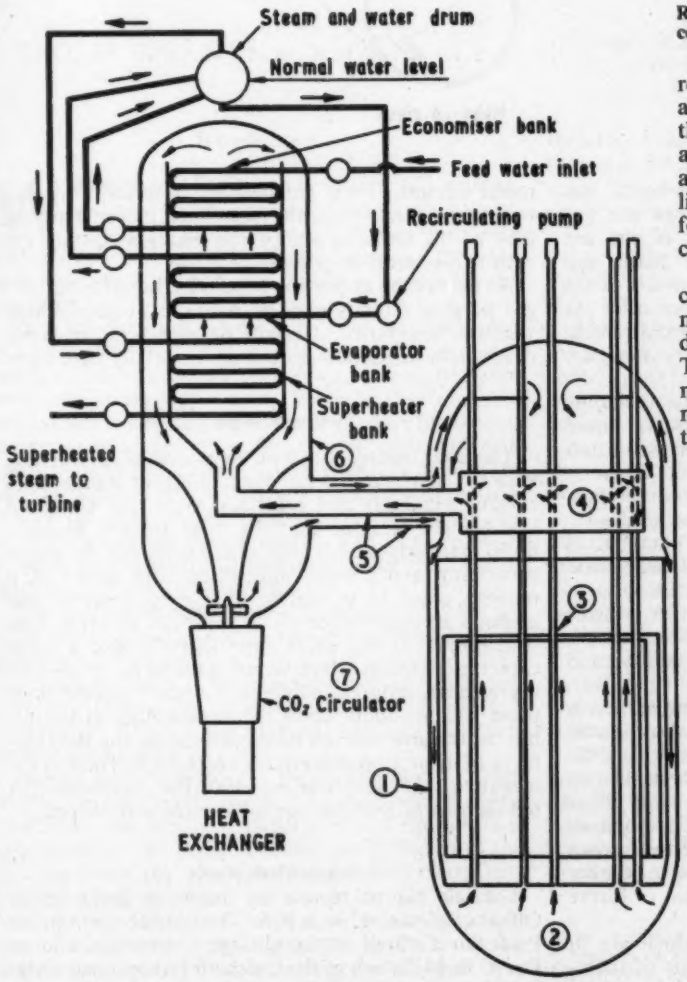
No account of welding and the nuclear power programme would be complete without reference to the fast fission breeder type of reactor. The present assessment of this system indicates that it will give the lowest cost of electricity of all types that have been considered. A reactor experiment is at present being conducted at Dounreay. Although this will probably be followed in the next few years by a prototype, the development problems are so great that it is unlikely that this reactor can be considered industrially before the 1970's.

Fuel element requirements

The present fuel element is different from that of the gas cooled reactor and consists essentially of annular fuel components contained within a niobium outer tube and a vanadium inner one. The inner vanadium tube fits into an end plug of niobium and is free to move. No welding is carried out on the vanadium. At the top and bottom of the fuel element the outer niobium tube is joined by butt welds to end plugs which accommodate the inner vanadium tube (Fig. 8).

Niobium is readily weldable by the inert-gas tungsten-arc process provided that contamination of the weld pool by oxygen, nitrogen, and hydrogen is avoided. To ensure this the complete operation is carried out in an enclosure which is evacuated and then filled with pure argon to a pressure sufficient to maintain a stable arc.

On completion, these fuel elements are filled with sodium to ensure good heat transfer, and the can is vented to allow expansion of the sodium and escape of gaseous fission products. Proof against leakage is not required, but after every six production welds (*i.e.*, after every three fuel elements) a test weld is subjected to a tensile test. The breaking load is in the range 0·9 to 1·1 tons, corresponding to a stress of 20–25 tons/sq.in. in the outer niobium tube.

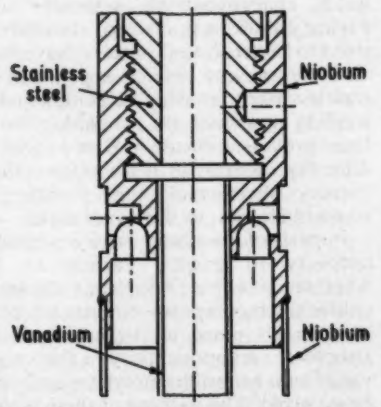


7—*Advanced gas-cooled reactor*

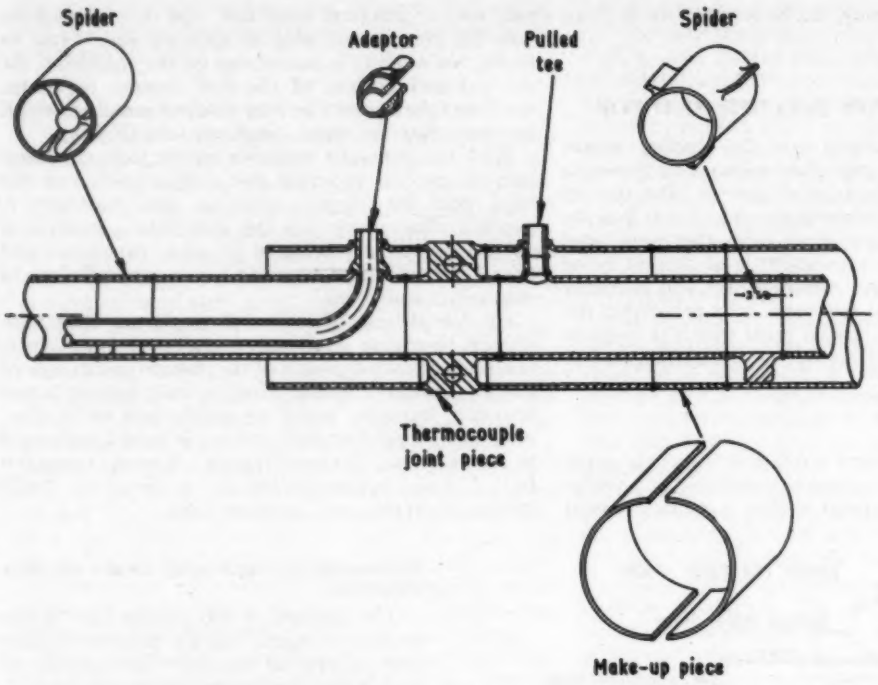
Requirements for liquid metal circuits and other components

The coolant of the present fast fission reactor is liquid sodium–potassium alloy and because of the high flammability of these metals in air, an extremely high standard of welding is necessary to avoid leakage at all times. All welds in contact with the liquid metal must be X-rayed and therefore for satisfactory radiography only butt welds can be used.

Although all welded joints in the coolant circuit are radiographed, very fine defects might escape detection unless suitably disposed in the direction of the X-ray beam. The possibility exists that such a defect might extend right through the wall thickness of the heat exchanger. Gaseous leak tests are therefore carried out in addition to



8—*Butt weld at end of fast reactor fuel element*



9—Use of butt welds for liquid sodium plant to permit radiography

radiography. During fabrication nitrous oxide was used, while for site assembly a halogen leak test was used because of the greater portability of the test equipment. On completion, the whole plant was subject to a nitrogen pressure test, the outside of the welds being painted with a detergent. Even after the previous intensive inspection a few leaks were found at this stage. Although only very small, they were unacceptable in a plant of this kind.

Welding is involved in all components, including heat exchangers, pumps, measuring devices, and liquid metal circuits in general, but the invariable use of butt welds challenges the ingenuity of the designer. Figure 9 shows a number of examples where modifications to conventional practice have been necessary.

In the present reactor experiment the circuit is of stainless steel, for which accepted and well established welding processes are available. The specification of these processes for use in liquid metal circuits presents difficulty, notably in defining the effect of inclusions, cavities, micro-cracks, under-cutting, oxidation, and so on in relation to the liquid metal.

In parts of the circuit there is a compelling economic necessity to employ carbon or low-alloy steels wherever possible. This leads straight away to difficulties from composite circuits where such steels may have to be joined to the austenitic type. For other structural components of the reactor, niobium, vanadium, tungsten, molybdenum, or zirconium may be involved. The welding of these metals to each other and possibly to steels may be a requirement of future reactor designs.

Non-destructive methods of testing, which are in general applied to pressure vessels, do not of themselves give the sort of guarantee required for liquid metal circuits. They are needed, however, for containment vessels, where the main duty, if ever required, may be the arduous one of containing a sodium fire with fission product release.

As part of the inspection procedure liquid metal and gas purging circuits may be required to satisfy high vacuum standards. Urgent development of non-destructive inspection procedures is vitally necessary.

Conclusions

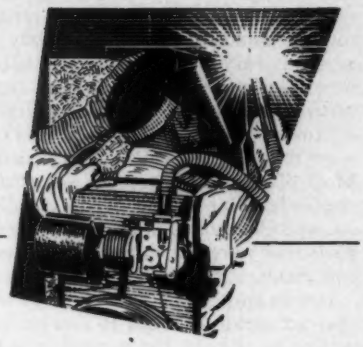
The civil reactors to date have not called for any basically new material but demand higher standards of existing materials and practices. With the Advanced Gas-Cooled Reactor and the Fast Fission Breeder, novel materials are introduced accompanied, by major processing and joining difficulties. The safety of a nuclear plant is a matter of public concern and assumes an importance which overrides all other considerations. At the same time there is also a large capital investment which would have to be written off if premature failure took place. Research and development will no doubt solve the outstanding problems, but the rewards will not be reaped unless the development of inspection procedures keeps pace. There is the incentive, the justification, and the challenge for industry to provide the answers to these problems.

Acknowledgments

I should like to express my thanks to those Senior Officers of the U.K.A.E.A., Industrial Group, to Professor Cottrell of Cambridge University, and to Dr. A. B. McIntosh of the Culcheth Laboratories, who have helped me in the preparation of this paper.

Welding

TECHNOLOGY



Welding from the Viewpoint of Lloyd's Register of Shipping

By G. M. Boyd, M.I.N.A., A.M.I.STRUCT.E.

The paper is a historical survey of the attitude of the Society towards welding during the past fifty years.

Introduction

WELDING, and in particular electric arc welding, is one of those technical developments which has spread with little regard to economics, and with a speed that has sometimes outstripped the complete solution of many of the problems to which it has given rise.

Such developments—and others can be cited, such as the introduction of iron, and later steel, into ship construction—have met with both enthusiastic support and severe opposition. Their real nature, *i.e.*, revolutions in engineering technology, were recognized in the early stages by a few people endowed with unusual forward vision. These pioneers tended to become known as 'enthusiasts' whereas they were mostly far from such. They merely recognized the inevitability of the developments, and were resolved to be in the vanguard of progress rather than in the rearguard of the opposing forces.

The Authorities faced with the responsibility for the reliability of structures and machines affected by such technical developments are in a particularly difficult intermediate position, and it is the object of this survey to review the past and present attitudes of one of these authorities, Lloyd's Register of Shipping, towards electric arc welding.

During its long history, Lloyd's Register of Shipping has encountered many similar developments, and its attitude has been consistently one of cautious encouragement. The Society, rather oddly, is part of the very industries for which it sets standards of quality, and is indeed governed by the industries which it governs. This being so, its objects are primarily to help the industries to progress by means of the most advanced methods, while maintaining an adequate and uniform standard of reliability. The Society must therefore keep, not only abreast of, but in advance of, technical developments, so that at the earliest possible stage the necessary criteria of reliability can be established.

Historical

Although electric arc welding, as a method for joining metals, was known in the latter part of the 19th century, its first real emergence as a factor in engineering technology occurred in the first decade of the present century, when it was introduced mainly for repairs. The first case in the records of the Society was the repair of the combustion chamber seams of a boiler, in March 1906. The repairs were found to be satisfactory in several subsequent examinations. During the next five years a total of 160 boilers in ships classed with the Society were similarly repaired. In the same period there were twenty instances of arc welded repairs to hull structures, mainly for the correction of flaws and cracks found in sternframes and rudders. The experience with these hull repairs was varied, most being satisfactory, but in some instances the sternframe or rudder affected had eventually to be renewed.

The first authoritative suggestion that welding could be used for the repair of boilers was made by Mr. H. Ruck-Keene of Lloyd's Register;[1,2] he later became Chief Engineer Surveyor. In August 1911, a circular was issued giving the conditions upon which arc welding could be accepted, the essence of which was

Paper presented to the North Eastern (Tyneside) Branch of the Institute on 8th January 1959.

Mr. Boyd is Technical Records Officer at Lloyd's Register of Shipping.

that the time had not arrived for the general and unqualified approval of welded repairs to important forgings and castings, and that they would be regarded as temporary, subject to examination from time to time. This indeed applies to some types of forgings and castings today. The use of welding for other repairs continued to extend, and in 1917 about 154 cases were reported, of which 140 were repairs to boilers. Most of these early repairs were made with bare wire electrodes, but these were rapidly superseded by covered electrodes. The boiler repairs were confined to parts that were not subjected to direct tension, such as end plates, combustion chambers, and furnaces.

It was only in the early days of the 1st World War that arc welding began to emerge as a possible competitor to riveting for normal structural connections, and in 1917 extensive experiments were carried out in the U.S.A. with the collaboration of the Society, and also by the British Admiralty at Portsmouth. These tests were mainly concerned with the ultimate tensile strength of welded joints.

In the following year the Society carried out, at the Birkenhead works of Cammell, Laird & Co. Ltd., an even more comprehensive series of tests, including tensile tests in a 300 ton machine, bending tests under alternating loads, and impact tests. The details of these tests were given by Sir Westcott Abell, then Chief Ship Surveyor to Lloyd's Register.[3] As a result of these tests it was decided that if the scantlings and arrangements were generally equivalent to those required for a riveted ship, and that the system and procedure were approved, welding could be accepted for classed ships, subject to the notations "Experimental" and "Subject to Annual Survey." There was also a special notation to the effect that electric welding was used in the construction.

In August 1918 the Committee of Lloyd's Register issued *Tentative Regulations for the Application of Electric Arc Welding to Ships*, the first rules of this kind ever to be framed.[2] Later the same year, the Society issued the first rules for the testing required for the approval of electrodes.

The first all-welded ship, so far as is known, was a 125 ft long barge, built at Richborough in 1918 for the War Office, and mentioned in a paper by J. Caldwell and H. B. Sayers.[4, 5] The first all-welded seagoing ship was the *Fullagar*, a 150 ft coaster built by Cammell, Laird in 1920 to Lloyd's Class with the notations "Electrically welded, subject to annual survey, experimental."[6] This ship had a long and eventful career of strenuous service with several changes of name and ownership. She went aground several times, and on each occasion subsequent examination revealed severe indentations but no ill effects to the welded joints, and there was no leakage. A collision caused serious indentations of the side shell in way of the hold and engine room, but the welding remained intact. In June 1924 the ship suffered serious grounding in the Mersey, and this set up the bottom 11 in. from bilge to bilge over a length of 70 ft. The damage was so severe that the Underwriters agreed to a total loss. Nevertheless, the ship refloated on the next tide and proceeded to Belfast, and later to Leith under her own power. There was no leakage, and the damage was repaired under the Society's supervision by forcing the deformed parts back into position with jacks and shores. After these repairs the Lloyd's Class was maintained without further special notation. She later travelled across the Atlantic, through the Panama Canal to the Pacific Coast, where she was employed in the cement trade, and subsequently suffered severe collision damage forward; although the plates were holed and torn, the welds remained unbroken. After over 11 years' service, the period between surveys was extended to two years, and the notation "Experimental" was deleted in 1933. After 17 years' service, the ship was finally sunk by collision in thick fog in the River Plate on 31st August 1937.

This ship may be regarded as a valid full-scale experiment in welded construction, under specially arduous conditions, and was an unqualified success. It was generally agreed that her structural behaviour was greatly superior to that to be expected of a comparable riveted ship under similar conditions. The resulting stimulus to welding was enormous, and the use of the process was greatly accelerated.

There were set-backs, of course. One of these was the case of the *Joseph Medill*, an all-welded ship for service in the Great Lakes, which disappeared on her first voyage across the Atlantic in August 1935. The reasons for this loss were never discovered, but the Court of Enquiry found that the most probable cause was impact with ice. It is of interest that her sister ship, the *Franquelin*, built in 1936, and also classed with Lloyd's Register, is still afloat.

Unquestionably, the loss of the *Joseph Medill* was a severe setback to the progress of welding for ship construction, even though the Court of Enquiry specifically eliminated faulty construction as a possible cause. Nevertheless, the faith of the pioneers was unshaken. In this connection it is of interest to quote a remark by a Mr. C. S. Lillicrap of the Admiralty. During the discussion of one of his papers[7] in November 1935 (soon after the loss of the *Joseph Medill*) he said:

> "*Any attempt made to put back the clock in the application of welding to shipbuilding is doomed, and rightly doomed, to failure.*"[8]

The validity of this prediction has been proved by the events of the ensuing years.

The author of this classical remark later became Sir Charles S. Lillicrap, K.C.B., M.B.E., D.Sc., Director of Naval Construction at the Admiralty, President of the Institute of Welding and of the British Welding Research Association. In case Sir Charles may be labelled as a "welding enthusiast" who happened to be fortunate, it is as well to quote the following passage from his 1935 paper:[7]

> "*And may I say that in my opinion there is no greater danger to the future of welding than the unbalanced enthusiast, the man who would weld everything and everywhere irrespective of the suitability of the job for welding and without appreciation of the possible risks involved. At the other end of the scale, there are those whose attitude is the time-honoured one that practice must not get in advance of scientific research, and that a new process must not be used until we know all about it. Both these attitudes are to be strongly deprecated, and the best way is to progress steadily with due caution.*"

During the thirties, the Admiralty were more active than most other authorities in the promotion of welding, largely due to the vision and dynamic personality of Mr. Lillicrap, and under the incentive of the Washington Conference to reduce weight. The Merchant Shipbuilding industry was somewhat slower in responding, probably because of reluctance to make the necessary radical changes in layout and skills required before the future of welding was fully established.

In May 1935 welding history was made by a Symposium on the Welding of Iron and Steel organized by The Iron and Steel Institute[19] in collaboration with fifteen other Technical Institutions. The symposium comprised 150 papers, covering all aspects of welding at the time, including several contributions by members of Lloyd's Register Technical Staff.

The Second World War

The progress of welding attained its highest rate, and also suffered some of its most severe setbacks during the Second World War. The enormous programme of emergency ship construction, which was carried out mainly in the U.S.A., was based almost entirely upon welded construction. The main reasons for this were that the shipbuilding facilities had to be vastly and rapidly expanded. Welders could be trained more rapidly than riveters, and welding was more adaptable for high-speed mass production, involving extensive prefabrication. New shipyards were built, and old ones reorganized, on this basis.

The first stage of the American emergency programme commenced in 1941–42 before the U.S.A. entered the War, and consisted in a British order for 60 identical 10,000 tons deadweight cargo ships based on a design developed on the North East Coast, *i.e.*, the "North Sands" design of Messrs. J. L. Thompson Ltd., of Sunderland. These 60 ships, known as the "Ocean" ships, were built under Lloyd's survey and gave yeoman service during the war and subsequently.[9]

Lloyd's Register of Shipping collaborated actively in this programme, both during the early negotiations and conversion of the design details to welding, and during the construction. Several surveyors, specially selected for their experience with welding, were sent to the U.S.A. to supervise the construction.

After the U.S.A. entered the War, the Emergency Shipbuilding Programme was stepped up, and the "Ocean" design, further slightly modified, was adopted as the basis for the famous "Liberty" and "Sam" ships. The same basic design was used for the "Fort" and "Park" ships built in Canada, though the construction of these ships was on more conventional lines, incorporating some riveting. A large tanker building programme was also launched in the U.S.A., based mainly on a 16,000 tons deadweight welded design, of which a considerable number had been previously built by the Sun Shipbuilding Co., of Chester, Pa.

Lloyd's Register, and the British Corporation Register, which were later amalgamated, continued to collaborate with the American Bureau of Shipping in these enormous programmes. The rate of production was amazing. Altogether, 5,777 ships with a total gross tonnage of nearly 40 millions were produced under the U.S. Maritime Commission emergency programme during the period 1939 to 1945. Of these, 2,748 were "Liberty" ships and 705 were tankers.[10] These were in addition to the "Ocean", "Fort" and "Park" ships already mentioned.

In general, the welded ships built during the War gave very satisfactory service, and indeed the performance under extremely arduous conditions was quite noteworthy. A. F. Davis[11] gives many examples of these ships surviving heavy enemy action. On the other hand some of the ships (quite a small proportion) suffered serious structural failures, some breaking in two, and others fracturing extensively. The failures were largely independent of enemy action, and gave rise to very extensive research work, to which reference will be made later.[12]

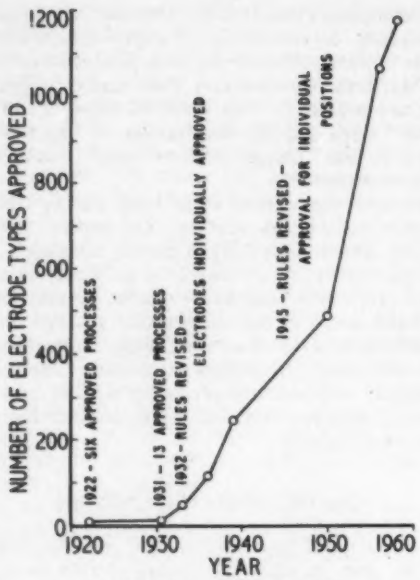
General Attitude of the Society

It is a general principle of the Society that the reliability and standard of quality should be judged from the end product, rather than from the method by which this is obtained. Thus, a weld is judged by its final quality rather than by the process or procedure by which it is made. For this reason, welding processes are not 'approved' as such, but when a new method is proposed, the surveyors collaborate with the manufacturers and users to ensure that satisfactory welds can be produced by this means. This general principle may of course be varied in individual cases. It is a corollary of this principle that certificates of competence are not issued to individual welders, the criterion being that the welder should in fact produce acceptable welds, rather than that he should be capable of doing so.

Naturally, the standards required, and the rigour of the control exercised over them, varies with the application. Clearly, a less severe standard may be tolerated for a secondary connection in a ship than for a high-pressure boiler drum or for an atomic reactor vessel.

Lloyd's Register is a world-wide organization, with offices in 177 ports distributed in 51 countries—almost every maritime country in the world. The reports from the surveyors form an enormous fund of factual knowledge on the construction of ships and their machinery, and on their behaviour in service. The reports are conserved and extensively cross-indexed by a special method, so that the relevant information on any particular aspect can be readily extracted.

It is a policy of the Society that the experience gained by the surveyors in their contacts with building, repairs, and maintenance all over the world should, in so far as is practicable, be made available to the industry generally. This is done by direct contacts and discussions with the industry, and by means of individual papers to the professional Institutions, articles in the technical press, and by the Society's own publications. The bibliography at the end of the paper gives some of the more important recent contributions by members of the Society's staff, on subjects relating to welding. Many other subjects have also been covered



1—Lloyd's approved electrodes

in a similar manner, but the list is too long to be given here.

Shipbuilding

The Society's Rules for welded ship construction are framed in such a way as to allow considerable latitude for development, but the main underlying principles are firmly emphasized. These are, briefly, aimed at continuity, soundness of the welds, full penetration, and the avoidance of notches and discontinuities of all kinds. Tables of dimensions are given for fillet welds, intermittent welds (where permitted) and "scalloping". It is required that the welding operators are to be proficient in the type of work on which they are engaged, that the work be adequately supervised, and that the welding equipment be suitable and efficiently maintained. The procedures are to be established by sample tests under similar conditions to those that will obtain during construction. The electrodes are to be approved and shown to be suitable for the proposed purpose. The Society issues an annual list of approved electrodes, which at present includes over 1,000 types manufactured in almost every country. Some indication of the growth of welding in the world is given in Fig. 1, which shows the number of types of electrode on the approved list over the period since 1922. The Rules specify the tests, etc., required for the approval of electrodes. These include tensile, shear, and cruciform tests on actual welded specimens, tensile, bend, and impact tests on all-weld-metal specimens, and chemical analysis of the weld metal. The Rules require that the manufacturers' plant and methods should ensure uniformity, and provision is made for periodical inspection of the makers' works, as well as check tests to ensure that the standard is maintained.

Marine Engineering

The problems that arise in the welding of marine engines, boilers, piping, pressure vessels, gearing, electrical machines, and the like, though having some basic principles in common with the structural welding of ships' hulls, are complicated by additional factors such as fatigue, high temperatures, special materials such as alloy steels, the need for high dimensional stability, accuracy, and the need to guard against the potentially serious results of even minor failures. Many such problems have been dealt with and overcome in the Society's experience, and have been recorded in numerous papers by members of the Society's staff. Noteworthy contributions have, in particular, been made by Dr. S. F. Dorey, F.R.S. and Mr. H. N. Pemberton, past and present Chief Engineer Surveyors, respectively (see Bibliography).

The Rules give specific requirements for the welding of such engineering components as boilers and pressure vessels, gear wheels, gear cases, piping, diesel engine bedplates, columns and entablatures. The well known Chapter J of the Rules (Welded Pressure Vessels) is widely recognized as a compendium of good practice, and in particular Section 19 of that Chapter, in which the requirements for Class 1 fusion welding are set out, is in effect a world standard of quality. For any firm to qualify for inclusion in the list of those recognized by the Society for the manufacture of Class 1 fusion welded pressure vessels, their works must be inspected by a Surveyor, and a special series of tests including the manufacture of a full-size pressure vessel must be made to demonstrate that the high standards required by the Rules can be consistently achieved. There are at present (1957 Rules) only 132 firms so qualified in the world, 36 of which are in the United Kingdom, and the remainder in 16 other countries. The routine tests for Class 1 pressure vessels made by approved firms are severe, and include tensile, bend, and macro-etching tests on specimens welded during, and as a continuation of, the actual welding of the vessel, X-ray photographs of the entire length of each welded seam, and hydraulic pressure testing of the completed vessel. Heat-treatment of the complete vessel is also required.

The Society's requirements for the heat-treatment of some pressure vessels and other engineering components are widely regarded as sound practice, though on occasions their necessity in individual applications has been scrutinised, and the question is often asked whether the heat-treatment is intended to relieve residual stresses or to improve the metallurgical structure in the vicinity of the welds. The answer is that both benefits result from judicious heat-treatment, and particularly the treatment known as 'stress relieving'. *i.e.*, heating uniformly to 600–650°C. and cooling slowly in still air. It is well established that such heat-treatment considerably reduces the risk of cracking, and is beneficial in conserving dimensional stability when the parts are to be machined, as for example gear cases, gear wheels, entablatures and bedplates.

Arc welding is now used in a wide variety of marine engineering components, and its use is extending as techniques are improved, and as experience is accumulated.

Land Engineering

The scope of the Society's activites has for many years embraced many structures and engineering

components for essentially non-marine applications, and of recent years these have grown to such an extent that a special department, Lloyd's Register Land Division, has been formed to deal with them.

The basic principles that apply to marine structures and engineering clearly apply with equal force to their counterparts ashore, and these principles have been outlined. The objective of setting standards of good practice and reliability, with due regard to practical and economic conditions, applies throughout.

In general, the Society's Land Division deals with the "high grade" group of installations, such as power stations, hydro-electric plant, penstocks, wind tunnels, oil refineries, pipelines, and atomic power plants. The high standards of quality which are regarded as necessary and practical in marine work are applied for such land installations, with such additions as may be prudent or necessary according to the special circumstances of each case.

In some applications the necessary standards of reliability are very exacting, and it has been one of the most important functions of the Society to develop practical rules which will at the same time meet the severe requirements and yet be capable of implementation under practical conditions. This applies with special force to welding, which holds a key position in relation to safety.

The development of atomic power, which, as is well known, depends greatly on welding of a specially high standard, is taking place at present mainly in land applications, although its application to the propulsion of merchant ships is not far distant. Atomic installations entail special problems which are not of such immediate urgency in most other spheres, such as the very large sizes and thicknesses of some of the pressure vessels, the high degree of dimensional accuracy required, and the effects of irradiation. Further difficulties arise from the fact that much of the fabrication, welding, X-ray, stress-relief and pressure testing must be done on site, often in remote places.

The primary circuits, *i.e.*, reactors, heat exchangers, gas ducting, etc., at Calder Hall, Chapel Cross, Hunterston, and Bradwell, as well as sealed buildings such as the Dounreay sphere, and cylindrical vessels housing research reactors at Harwell and Dounreay, have been, or are being, built under the Society's survey.

It is not proposed here to go into details of the many welding problems involved in these land applications—this has been efficiently done in other papers[13–17]—but merely to indicate that the Society is devoting considerable and effective efforts towards their solution. The field is developing from day to day, and Lloyd's Register is in the forefront of these developments.

Research

Welding, like many other radical developments in technology, has brought its own particular problems, which have and are engaging the attention of many research workers and practical engineers. Among those problems may be cited those of residual stress, brittle fracture, fatigue, distortion, and the incidence of defects such as porosity, slag inclusions, and cracking. Some of these problems are, of course, not peculiar to welding, but they have come into prominence and urgency due to its widespread use.

Lloyd's Register has taken an active part in the research arising from these problems, by direct participation in the work of Committees, such as those of the Research Associations, the Admiralty, the British Standards Institution, and their counterparts in other countries. The Society has been able to contribute factual data from actual experience, and has of recent years carried out research work at its own Laboratories at Crawley in Sussex (Fig. 2).

It is not the purpose of this article to go into the details of the vast research on these problems, since these have been dealt with in several of the papers referred to in the bibliography. A review of this kind would, however, be incomplete without some reference to one of the most pressing and intractable of the problems—brittle fracture.

The serious fractures which occurred in welded ships, as mentioned earlier, were found to be brittle in character, a fact which conflicted with the accepted ductile nature of shipbuilding steel as shown by the normal tensile and bend tests. Even more surprising was the fact that samples of the steel and weld metal removed from the vicinity of the fractures showed adequate ductility, and compliance with the Rule requirements. The phenomenon could not be attributed to simple lack of strength, since the scantlings of the welded ships were not less than those of comparable riveted ships which had been successful. These and other apparent anomalies led to vigorous research in many countries, particularly in the U.S.A. and the U.K., which were most directly affected.

2—*Research Laboratory at Crawley*

Although many possible explanations and theories were investigated, the solution remained elusive and the problem was found to be far from simple. Indeed, it cannot be said to be fully solved even today.

One outcome of the work was, however, that it became apparent that the steel to be used in welded structures, and particularly in ships, required a closer control than was found adequate for riveted ships. Lloyd's Register took an active, and possibly leading part in the research on this point, and has made important amendments in its Rules to meet the situation.[18] The subject is, however, highly controversial even today, and much still remains to be learned about it.

The Future

It is now widely accepted that welding has come to stay, and indeed that it is essential to the present state of advanced engineering techniques. Many of the engineering and structural components in daily use today would be inconceivable without the use of welding. It can equally be predicted with confidence that riveting is obsolescent. In the circumstances the only course open is to press forward, and in so far as is possible to anticipate and solve the problems which this advance inevitably entails. Some new problems are already emerging, particularly in the fields of gas turbines, rocketry and atomic energy. Lloyd's Register is fully alive to them and, in accordance with the general policies outlined here, has already come to grips with them.

Prophesy is a dangerous thing at any time, but it can be predicted with confidence that the Society will meet, and take in its stride as hitherto, the problems which must accompany progress.

Acknowledgments

The author wishes to acknowledge the valuable help received from colleagues and others in the preparation of this brief survey.

BIBLIOGRAPHY AND REFERENCES

(Items marked * are by past or present members of Lloyd's Register Technical Staff)

1. *H. Ruck-Keene: "New methods of effecting boiler repairs." *Trans. Inst. Mar. E.*, 1907–8, vol. 19.
2. *H. N. Pemberton: "Welding in marine engineering construction" (Thomas Lowe Gray Lecture), *Proc. (A) Inst. Mech. E.*, 1953, vol. 167, p. 48.
3. *Sir Westcott Abell: "Experiments on the application of electric welding to large structures", *Proc. Inst. Civil Eng.*, 1918–19, vol. CCVIII, Pt. II, p. 104.
4. J. Caldwell and H. B. Sayers: "Electric welding developments in Great Britain and the U.S.A." *Proc. Inst. Civ. Eng.*, 1918–19, vol. CCVIII, Pt. II, p. 69.
5. J. Caldwell: *Trans. Inst. Eng. Ship. Scotland*, 1917–18, vol. LXI, p. 262 (Discussion).
6. *The Shipping World*, 1931, 22nd April.
7. C. S. Lillicrap: "Welding as applied to shipbuilding", *Welding Industry*, 1935, Nov., p. 360.
8. C. S. Lillicrap: "Future technique in shipbuilding", *Trans. Inst. Welding*, 1945, Feb., p. 12.
9. R. C. Thompson and H. Hunter: "The British merchant shipbuilding programme in North America" (Andrew Laing Lecture), *Trans. N.E. Coast Inst. Eng. Ship.*, 1942–43, vol. 59, p. 61.
10. F. C. Lane *et al.*: "Ships for Victory", The John Hopkins Press, Baltimore, 1951.
11. A. F. Davis: "The saga of welded ships", *Welding J.*, 1945, Jan., p. 80.
12. "The Design and Methods of Construction of Welded Steel Merchant Vessels", U.S. Govt. Printing Office, Washington, 1947.
13. *Weld. Met. Fab.*, 1956, Nov., p. 384.
14. *E. Crossley: "Inspection of large site-built pressure vessels" *Weld. Met. Fab.*, 1956, Nov., p. 393.
15. J. F. Lancaster: "The welding of aluminium tanks for experimental reactors", *Brit. Welding J.*, 1957, vol. 4, p. 354.
16. S. H. Griffiths: "Welding of heat exchanger elements in a fast breeder reactor", *Brit. Welding J.*, 1957, vol. 4, p. 403.
17. J. A. Lucey *et al.*: "Some welding developments applicable to the fabrication of heavy pressure vessels for nuclear power stations", *Brit. Welding J.*, 1957, vol. 4, p. 449.
18. *J. Hodgson and G. M. Boyd: "Brittle fracture in welded ships", *Trans. Inst. Nav. Arch.*, 1958.
19. "Symposium on the Welding of Iron and Steel", The Iron and Steel Institute, vol. II, 1935.
20. *R. B. Shepheard: "Recent developments in British merchant shipbuilding", *Trans. N.E. Coast Inst. Eng. Ship.*, 1944.
21. *R. B. Shepheard: "Developments in British merchant shipbuilding", *Soc. Roy. Belge des Ingenieurs et des Industrials*, Brussels, 1946.
22. *R. B. Shepheard: "Ship structures—a century of progress," *Trans. Inst. Nav. Arch.*, 1951.
23. *R. B. Shepheard: "Aspects of research in British merchant shipbuilding" (Adams Lecture). *Welding J.*, 1954, Jan., p. 23.
24. *F. C. Cocks: "General review of welding in shipbuilding", Conference on Welded Structures, H.M.S.O., 1954.
25. *S. F. Dorey: "Inspection and testing procedure, with special reference to welded pressure vessels and engine entablatures" (see ref. 19 p. 745).
26. *S. F. Dorey: "The welding of pressure vessels". *J. Inst. Pet. Tech.*, 1936, Feb.
27. *S. F. Dorey: "Welded joints in pressure vessels", *Proc. Inst. Civil Eng.*, 1937.
28. *S. F. Dorey: "The welding of boiler drums and shells", *Trans. Liverpool Eng. Soc.*, 1939, March.
29. *S. F. Dorey: "Progress in marine engineering as influenced by the classification of ships" (Thomas Lowe Gray Lecture). *Proc. Inst. Mech. Eng.*, 1941.
30. *S. F. Dorey: "Ferrous materials in marine engineering" (Andrew Laing Lecture). *Trans. N.E. Coast Inst. Eng. Ship.*, 1955, Oct.
31. *S. F. Dorey: "Notes on the design stresses in class 1 welded pressure vessels," *Proc. Inst. Mech. Eng.*, 1946, vol. 154, p. 32.
32. *J. M. Murray: "A hundred years of Lloyd's Register Ship Rules", *Trans. Inst. Nav. Arch.*, 1955.
33. *H. N. Pemberton: "The testing of welds", *Trans. N.E. Coast Eng. Ship.*, 1942, vol. LVIII.
34. *H. N. Pemberton: "Welding in marine engineering construction". *Trans. Inst. Nav. Arch.*, 1939, p. 59.
35. *J. Turnbull: "Hull structures". *Trans. Inst. Eng. Ship. Scotland*, 1956–57, vol. 100, p. 301.
36. *H. R. Gibbs and G. M. Boyd: "Welded ship construction—records of common fractures and their causes", *Trans. Inst. Eng. Ship. Scotland*, 1958, Feb.
37. *G. M. Boyd: "Effects of residual stresses in welded structures", *Brit. Welding J.*, 1954, vol. 1, p. 560.
38. *G. M. Boyd: "Brittle fracture", *Trans. Manchester Assoc. Eng.*, 1957, Oct.

Effects of Hydrogen in Wrought Steel and in Ferrous Weld Metal

By P. D. Blake M.A., F.I.M., and W. I. Pumphrey, D.Sc.

A review is given of much of the recent published work on the effects of hydrogen upon the mechanical properties of steel and of weld metal, more particularly where figures for the hydrogen contents of the metal have been published or can be estimated. It is concluded that the effects of hydrogen upon steel and upon weld metal are qualitatively similar, the apparent differences being largely due to the inherent difference in the microstructure and composition of weld metal and wrought steel.

In considering the preparation of this paper the authors were motivated by the thought that a factual appraisal of the effects of hydrogen in wrought steel and in ferrous weld metal is long overdue. A critical summary has therefore been made of most of the more recent papers on the effect of hydrogen on the mechanical properties of wrought steel and on the mechanical and cracking properties of weld metal.

Effects on Impact Properties of Steel

Heat-treatment, structure, and composition

The effect of hydrogen appears to be inextricably bound up with that of the prior heat-treatment and general history of the steel. Although the factors involved have not been established precisely, it is known or suspected that the solubility, diffusivity, and embrittling effect of hydrogen vary in the different micro-constituents of steel, such as pearlite, bainite, martensite, austenite, iron carbides. A complicating factor is that the heat-treatment of steels can and generally does result in a lowering of their hydrogen content. Hence it is not always easy to state whether the result of a given heat-treatment is due to changes in the microstructure, or to a reduction of the overall hydrogen content, and it is usually true that the change in properties is due to both effects.

The work of Hobson and Sykes[1] on the low-alloy steels shows quite clearly how heat-treatment can affect embrittlement. A typical example is the effect of charging 4½ ml of hydrogen per 100 g, by high-pressure hydrogenation at 300°C., into a ¼% C–3¼% Cr–Mo low-alloy steel, after an initial isothermal treatment at 690°C. Impact tests upon this steel at room temperature revealed that it had only half the impact value of the hydrogen-free steel in the same conditions; the same steel, air-hardened from 920°C., tempered for 4 hr at 710°C. and hydrogenated at 300°C. in the range 0–4½ ml/100 g gave only a 10–15% reduction in the impact value for the maximum hydrogen content at testing temperatures of 20°–100°C.

As another example, the work of Stout, McGeady, Sun, Libsch, and Doan[2] also affords confirmation of the theory that the heat-treatment of a steel has an important effect upon its susceptibility to hydrogen embrittlement. In all the tests by these latter authors, Charpy V-notch specimens were charged by soaking in wet hydrogen at 1117°C. for 10 min, followed either by quenching directly into water or by air-cooling or oil-cooling to a black heat and subsequently quenching in water. This gave a range of cooling rates, and testing was carried out within 30 min of the final quenching using an Olsen impact-testing machine with a striking velocity of 16½ ft/sec. A hydrogen concentration of about 5 ml/100 g was found to lower the impact energy values of ¼% C–1% Mn-Mo low-alloy steel by about 15–20 ft-lb, compared with unhydrogenated specimens. The transition temperatures of the air-cooled specimens rose by about 30°F., but not those of the quenched steels. For the ¼% C–2–3%Mn low-alloy steel, a hydrogen content of about 5½ ml/100 g was without effect upon the transition temperatures of specimens quenched by any of the three methods, although the impact value was lowered by 5–8 ft-lb, as compared with the unhydrogenated steels. In a third series of tests on 1% Mn–Ti high-tensile steel, a

Paper presented at a Joint Meeting with The Iron and Steel Institute on 10th December 1958. The discussion on this and previously published papers on "Hydrogen in Weld Metal" will be printed in a later issue of the Journal.

The authors are with Murex Welding Processes Ltd.

hydrogen content of 6 ml/100 g gave only a slight reduction in the energy values or increase in the transition temperatures of any of the specimens tested as compared with unhydrogenated specimens.

This investigation shows that the effect of composition or alloy content is important in any study of hydrogen embrittlement, in addition to that of heat-treatment.

Method of charging the gas

Similar results and variations in the effects of hydrogen appear to be obtained regardless of whether the hydrogen is charged by high-pressure hydrogenation in the range 300°–350°C., as in the work of Hobson and Sykes,[1] or by cathodic charging as performed by them and Herres;[3] or by soaking in wet hydrogen at 1100°–1200°C. as in the work of Stout McGeady, *et al.*[2]

Cathodic charging gives a high hydrogen concentration in the outer skin of the steel, which should be distributed through the material by heat-treatment at 200°–300°C., if the results are to be made strictly comparable to those obtained with high-pressure hydrogenation or high-temperature soaking.

Type of impact test

In general, the notched-impact tensile test appears to be less sensitive to effects due to hydrogen embrittlement than the ordinary Charpy type of impact test. More information is obtainable from the former type of test, however, and generally hydrogen is found to reduce the ductility of steels.

As an example, Hobson and Hewitt[5] found that in impact tensile tests on ¼% C–3¼% Cr–Mo low-alloy steels, hydrogen in the range 0–10 ml/100 g was virtually without effect upon the energy values or the elongation at 20°C.; however, the reduction in area fell about 10%. Yet Hobson and Sykes[1] found some quite striking examples of hydrogen embrittlement in the ordinary Charpy test using the same type and composition of steel. On the other hand, Herres[3] found that in testing cathodically charged ¼% C–1¼% Mn–Cr–Ni–Mo low-alloy steels by the notched-impact tensile test at 20°C., in the 'as-quenched' condition, hydrogen caused the reduction in area to fall from 40% for the hydrogen-free steel to 10–20% for the hydrogenated steel.

Need for replicate specimens

There has been a regrettable tendency, in this type of work on impact properties, for relatively few duplicate specimens to be tested, owing to the time-consuming nature of hydrogenation and heat-treatment. Nevertheless, it is highly desirable in a test known to exhibit such variability as the Charpy impact test that at least 8–12 and ideally about 20–30 replicate specimens should be heat-treated and hydrogen-charged in each state, so that the results are amenable to formal statistical analysis. This lack of replicate determinations leaves much to be desired in the existing published literature on hydrogen embrittlement in the impact test (or indeed in the tensile test), in which generally only 2–3 replicates have been tested. To some extent, as in weld-metal tests, the making of even eight replicates can become a very lengthy procedure which is not economically always practicable, but in tests on commercial steels there should be little difficulty. It is thought that this lack of replicate testing may account for some of the variability encountered in tests involving hydrogenation. If more data were available on which statistical analysis could be performed, more reliable interpretation of the results of investigations into the effects of microstructure and heat-treatment upon hydrogen embrittlement could be made.

Testing temperature

It has been a more or less generally accepted theory of hydrogen embrittlement that, as the testing temperature is lowered, the effects of hydrogen become proportionally reduced, but the temperature at which embrittlement ceases has been variously estimated at −160°C.[6] and −29°C.[7], although Baldwin and Brown[7] found that there was no embrittlement at any temperature at strain rates of 19,000 in./in./min. Hobson and Sykes,[1] in their work on 3% Cr–Mo low alloy steels at temperatures between −50°C. and +100°C., found that hydrogen lowered the energy values by about 10–20% at 100°C. but by as much as 50% or more at room temperature and −50°C. in many cases. On the other hand, 2½% Ni–Cr–Mo low-alloy steel was virtually unembrittled[1] by hydrogen (2–3 ml/100 g) at +20°, +50°, or +100°C., though the hydrogen-free steel had only a low impact energy. It is clear from this work that temperatures much lower than −50°C. are required to eliminate the effect of hydrogen on low-alloy steels.

It must also be borne in mind that, as found by Baldwin and Brown,[7] above a certain speed of impact testing, probably about 10,000 in./in./min, the speed of testing eliminates hydrogen embrittlement.

Properties at room temperature

As the results already quoted show, whether hydrogen has an embrittling effect at 20°C. depends very much on the composition, micro-structure, and prior heat-treatment of the specimen. In contrast to the results of Hobson and Sykes[1] many examples can be quoted in which hydrogen is without effect upon the impact properties. As further examples, Marshall, Garvey, and Llewellyn[4] found that hydrogen had a negligible effect upon the transition temperature of notched-bar impact specimens of wrought *A.I.S.I.* steel which had shown marked embrittlement in the tensile test. Again, Herres[3] found that hydrogen was without effect upon the Charpy energy levels at 20°C. of a ¼% C–1¼% Mn–Cr–Ni low-alloy steel; however, this example is somewhat ambiguous because much of the cathodically charged hydrogen may have been lost during a heat-treatment at 871°C., which was carried out after charging, followed by oil quenching.

Much systematic work in this field is required before any theory of the effects of hydrogen can be formulated; it can only be concluded that in the present state of knowledge it is impossible to predict in any given case whether hydrogen will have an embrittling effect or not.

Separate effects of diffusible and residual hydrogen

Virtually no systematic attempt at comparing the effects of hydrogen in the lattice with that in the voids

Table I

Approximately equivalent British and American welding electrode classes for mild steel

Electrode	*Type*	*Bonding Agent*	*BS.1719 (1951) Class*	*AWS Class*	*Position*
Cellulosic	Mild steel	Sodium silicate	E.110P, E.11XP	E.6010	a.c. all positions and d.c. electrode +ve
Cellulosic	Mild steel	Potassium silicate	E.111P, E.11XP	E.6011	
Rutile	Mild steel	Sodium silicate	E.212, 213, 215, 216, 217	E.6012	a.c. all positions
Rutile + easily ionisable materials	Mild steel	Sodium silicate	E.313 or 317	E.6013	d.c. electrode −ve
Lime	Mild and high tensile steel	Sodium silicate	E.610 or 61X	E.6015	Recommended d.c. electrode +ve. All positions
Lime	„ „	Potassium silicate	E.611 or E.61X	E.6016	„ „ „
Iron oxide	Mild steel		E.426 or E.426P, E.446, E.446P	E.6020	a.c. and d.c. either polarity. Flat and h-v positions
Iron oxide	Mild steel		E.436, E.436P, E.437, E.437P	E.6030	a.c. and d.c. either polarity. Flat position only.

in a metal has been published, yet work in this field could make a valuable contribution to any theory of hydrogen embrittlement.

Effects on Impact Properties of Weld Metal

It is known that weld metal is peculiar in many of its metallurgical features and, in the present state of knowledge, it is best dealt with in terms of the electrodes used in depositing it. A number of attempts have been made to assess the effects of hydrogen by using electrodes having different hydrogen contents in their coatings but, owing to the multiplicity of other metallurgical variables that are thereby introduced, these results are to a large extent invalidated. Nevertheless, it is useful to consider them in any review of the subject. Much of the work in this field also has been vitiated to some extent by failure to measure accurately even the total concentration of hydrogen involved, and recent work suggests that the effects of diffusible and residual hydrogen are different. For reference, the nearest equivalent British and American electrode classifications are given in Table I.

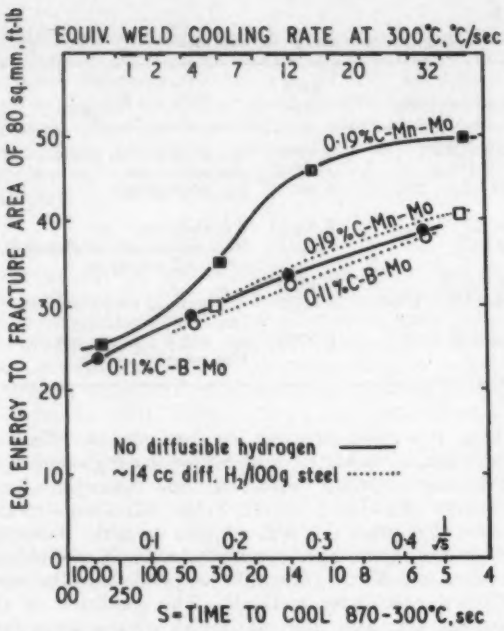
Stout, McGeady, *et al.*[2] investigated the effects of hydrogen by comparing different electrodes. They compared the effects of using the high-hydrogen AWS Class 6010 and the low-hydrogen basic electrodes in welding a ¼% C–2–3% Mn steel, a ¼% C–1% Cr–Mo–Ni low-alloy steel, and a 0·13% C–1% Mn–Ti high-tensile steel. The transition temperatures of Charpy V-notch specimens were the same for both deposits and roughly equal to that of the base steels, but the E.6010 weld deposits showed a much greater degree of scatter than the basic electrode deposits. The basic electrode deposits showed energy values roughly 5 ft-lb higher than for the E.6010 deposits, but this could well be accounted for by differences in the oxygen content of the welds. Cellulosic electrodes in general have higher oxygen contents than basic electrodes (*see* Table II) and increasing oxygen is known to impair the impact properties of iron and iron-carbon alloys.[8] Moreover, basic electrode deposits generally contain higher manganese contents than those from cellulosic electrodes and this element is known to lower the transition temperature of low-carbon steel.[9,10] Furthermore, basic electrode deposits also tend to have lower nitrogen contents than cellulosic ones and nitrogen is known to raise the transition temperature of iron.[10] Hence it is quite incorrect to investigate the effects of hydrogen in welds by comparing the high-hydrogen cellulosic deposits with the low-hydrogen basic deposits (BS.1719 Class I). Stout, McGeady, *et al.*[2] found that grain size was another variable, increased grain size in the base metals, as a result of welding, tending to raise the transition temperature of the heat-affected zone quite markedly. The condition of the coating, dry, wet, and oil-soaked, of the same electrodes had little effect on the transition temperatures of the deposits.

Very few experimental data are available on the effect of variations in the hydrogen content of a single electrode type upon the impact energy figures or transition temperatures. The results of Pellini and Eschbacher[11] on the drop-weight nil-ductility (DWT) transition temperatures of several AWS classes of electrodes are of some interest, in that they cover a wide range of hydrogen contents, but the large number of other metallurgical factors involved make it impossible to draw any conclusions as to the real effect of hydrogen on the DWT temperature. No hydrogen determinations were made in this work, but for the purpose of this review the results are arranged in Table II together with typical hydrogen and oxygen contents found by the authors for similar electrodes. The oxygen content is included here because it is

Table II

DWT transition temperatures and gas content for some AWS classes of electrodes

Electrode AWS Class	*DWT Transition Temperature*[11]	*Typical Gas Contents Hydrogen, ml/100 g*	*Oxygen, wt. %*
E.6010	−10° to +20°F.	15–20	0·06
E.6020–6030	+10° to +20°F.	5–8	0·14
E.7016	0° to 20°F.	1–5	0·03
E.6012	+30° to +40°F.	10–18	0·09
E.8016 (Ni)	−10° to −30°F.	1–3	0·03
E.10016 (Ni-Mo-V)	−50° to −70°F.		
E.12016 (Ni-Mo-V)	−60° to −70°F.		
E.10016 (stress relieved at 1150°F.)	+40° to +60°F.		
E.12016 (stress relieved at 1150°F.)	+80° to +90°F.		
E.10015 (Mn–Mo)	−60° to −70°F.		
E.10015 (stress relieved at 1150°F.)	−40°F.		



1—*Effect of variations in hydrogen content and cooling rate on impact tensile properties of B–Mo and Mn–Mo steels. Test temp. 20°C.: notched specimens; peak temp. 1250–1350°C. (After Cottrell[12])*

interrelated with the hydrogen content; in general, an increasing hydrogen or oxygen content would be expected to raise the transition temperature, although the effect of hydrogen is a little uncertain and the presence of other alloying elements, such as manganese, modifies the picture. The extremely low DWT temperature of the Mn–Mo electrode (E.10015) deposits can be related to the known effect of manganese[9, 10] in lowering the transition temperature, and when the low hydrogen content is combined with a low oxygen content, as with the EXX.016 classes, a low transition temperature occurs.

In the E.6010 class, with a high hydrogen and fairly low oxygen content, the two are in opposition and a moderately low transition temperature is found; this is also the case with the E.6020–6030 class of electrodes in which a low hydrogen content is combined with a high oxygen content. On the other hand, in the E.6012 group high hydrogen and oxygen contents are accompanied by a high transition temperature (+30° to +40°F.). Many other factors are involved here and there is a great need for reliable data from which most of the variables other than gas compositions have been eliminated.

Owing to the difficulty of obtaining unambiguous results in any investigation of the effects of hydrogen that is based on a comparison of different types of electrodes, it is necessary to consider alternative approaches. One method is to carry out investigations using the same electrode throughout but retaining different amounts of diffusible hydrogen in the weld by variations in the welding technique and/or in the cooling rate below about 250°–300°C., and also by machining some specimens under various low-temperature cooling media, such as liquid nitrogen or freezing acetone. This line of approach has been adopted by the authors in a paper to be published shortly. The other line of approach, which has been adopted by Cottrell[12] is to carry out all measurements on the base steels, which are cathodically charged and then subjected to a simulated welding cycle by means of high-frequency heating, followed by controlled gas or liquid quenching, to simulate the rapid cooling rates of welds. Cottrell charged all his specimens with hydrogen to a more or less uniform level of 14 ml/100 g, although even here the specimens tested for hydrogen could not be used in the tests. He found that this hydrogen level had no effect upon the impact tensile properties at room temperature of a 0·11% C–B–Mo steel over a wide range of simulated weld cooling rates, as shown in Fig. 1. On the other hand, it can be seen that over the same range of cooling rates there was a marked reduction in the energy to fracture of a hydrogenated 0·19% C–Mn–Mo steel, as compared with the unhydrogenated steel.

The work of Cottrell[12] appears to give unambiguous results for metal in the heat-affected zone, and further work on these lines should yield valuable information; on the other hand, it is more difficult to translate it into terms of welding practice because the inclusion content, grain size, and even microstructure of weld metal is so different from that of the base steels. The effect of even small amounts of alloying elements is seen to be an important factor in hydrogen embrittlement.

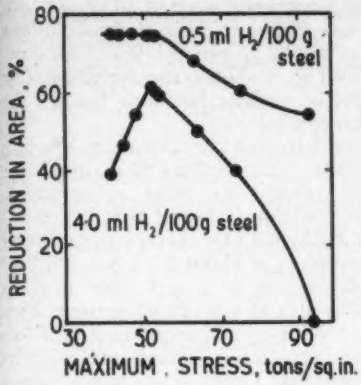
Effects on Tensile Properties of Steel

Heat-treatment and structure

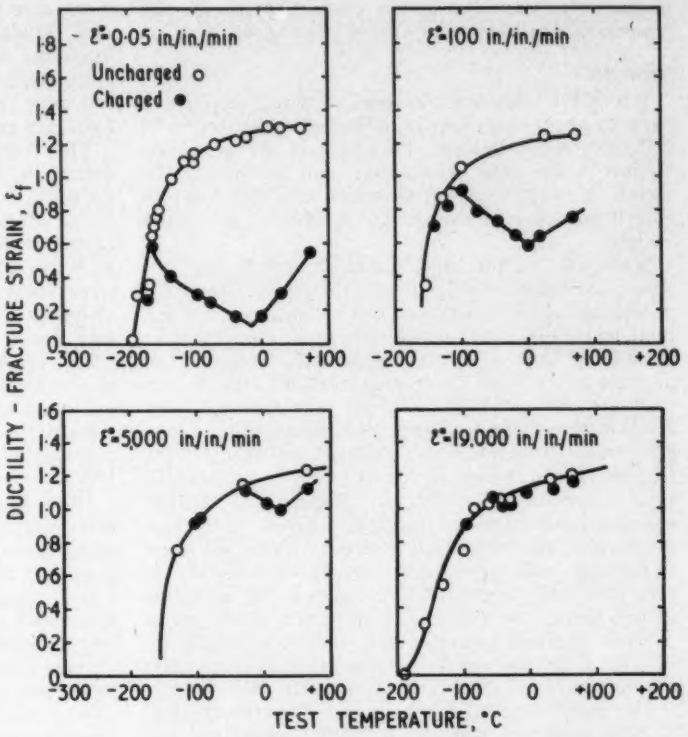
As with the impact properties, the effect of hydrogen varies considerably according to the prior thermal history of the steel. The work of Hobson and Sykes[1] provides an excellent example for ¼% C–3% Cr–Mo steels. Figure 2 shows how, at a given hydrogen level, the degree of hydrogen embrittlement is dependent on the nominal ultimate tensile strength produced in the steel by heat-treatment before hydrogenation. These results are of considerable interest because they indicate how the embrittling effects of hydrogen vary according to the microstructure produced by heat-treatment, there being an optimum treatment temperature and a corresponding ultimate tensile strength at which hydrogen embrittlement is minimized. This structure effect seems to indicate a promising line of research. The dependence of the degree of hydrogen embrittlement upon the prior heat-treatment is also apparent from the investigations of Hobson and Hewitt[5] and others.

Testing temperature

This is interdependent with the thermal history, the temperature of maximum embrittlement varying with differences in the prior heat-treatment of the steel. An interesting feature of some, but not all results, is the lack of any definite trend due to the temperature of testing. Hobson and Hewitt[5] in their work on ¼% C–3% Cr–Mo low-alloy steels have provided some examples of temperature effects. For example, when heat-treated to give maximum stresses of 41·5, 53, 67, and 98 tons/sq.in., the corresponding test temperatures

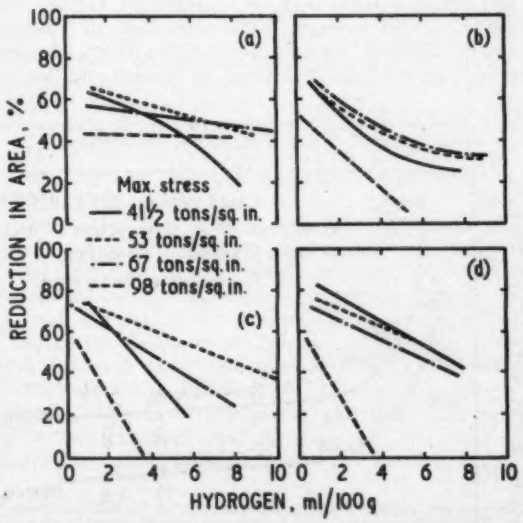


2—*Relationship between max. stress and reduction in area at two levels of hydrogen content for 3% Cr–Mo steel. (After Hobson and Sykes[1])*



4—*Effect of testing speed and temp. on degree of hydrogen embrittlement. (After Baldwin and Brown[7])*

at which maximum embrittlement occurred were 20°, −78°, and 20°C., and over the range 20°–100°C. respectively, as shown in Fig. 3. There is, however, a definite tendency, which is apparent from an inspection of the curves, for the degree of embrittlement to increase with increasing hydrogen content as the temperature rises from −196°C. to 20°C. For the steel having a nominal tensile strength of 98 tons/sq.in., the



3—*Effect of testing temp. on optimum max. stress for resisting hydrogen embrittlement of a 3% Cr–Mo steel:* (a) *−196°;* (b) *−78°;* (c) *20°;* (d) *100°C. (After Hobson and Hewitt[5])*

embrittlement actually does increase steadily with test temperature. Although Fig. 3 shows that there is less embrittlement at −196°C., the embrittlement is not by any means eliminated at this temperature.

The effect of testing temperature on the hydrogen embrittlement of carbon steels and of low-alloy steels is qualitatively similar, but has not been so thoroughly investigated for carbon as for low-alloy steels. Herres[3] found that the degree of embrittlement of cathodically charged steel was much less at −73°C. than at +21°C. Bastien and Anzou,[6] in contrast to the previously quoted results of Hobson and Hewitt[5] on low-alloy steels, found that a plain 0·15% carbon steel was not embrittled at all either at or below a test temperature of −160°C. or above +65°C.; however, the true rupture stress and percentage reduction in area were both reduced by an unspecified amount of hydrogen at temperatures between −70°C. and +18°C. Similar results were obtained by Baldwin and Brown[7] on SAE 1020 steel. Bastien and Anzou[6] also found that the temperature effect was reversible, in that specimens which were unembrittled at −160° again became brittle at above −110°C. They developed a theory that the immunity to embrittlement below −160°C. was due to the concentration of hydrogen adsorbed on the wall of the voids, changing from activated to physical adsorption, the latter being a much stronger type of adsorption leading to a decrease in the triaxial stresses. This would be a reversible process.

Baldwin and Brown[7] have shown that the effect of temperature upon embrittlement varies with the strain rate, the critical temperature below which embrittlement ceases being about −168°, −118°, and −29°C. at strain rates of 0·05, 1·00, and 5000 in./in./min,

respectively, there being no embrittlement at any temperature at 19,000 in./in./min (*see* Fig. 4).

Strain rate

It is well known that the speed of tensile testing can have an important effect in determining the degree of hydrogen embrittlement. However, in the preceding section, it has been shown that even at some of the speeds of impact testing hydrogen can still have an embrittling effect, although the results are somewhat variable.

Seabrook, Grant, and Carney[13] found that for specimens containing 6½ ml of hydrogen per 100 g, the percentage reduction in area was constant for loading rates up to ½ in./min. That hydrogen was less effective at loading rates of 5–10 in./min was shown by an increase in the reduction in area from 20% for the low loading rates to 35–40% for the high loading rates. Hydrogen was without effect upon the ductility of low-alloy steels in the notched-bar impact tensile test of the Charpy type at testing speeds of 13,600 in./min with a time to fracture of 0·001 sec.[5] Bastien and Anzou[6] found that for times to fracture of below 10 sec in a tensile test, the embrittling effects of hydrogen begin to decrease, although Hobson and Hewitt[5] found that they were still evident to some extent in the reduction in area figures for a time to fracture in a tensile test of 3–5 sec. At strain rates in excess of 5000 to 10,000 in./in./min, hydrogen was found to be without effect upon the ductility by a number of investigators.[6, 13, 14, 15]

The embrittling effect of hydrogen upon the ductility of mild steel as a function of both temperature and strain rate has been discussed by Baldwin and Brown.[7]

Ductility

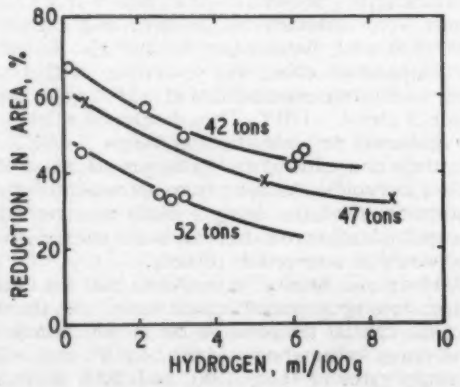
In general, there is a more or less inverse linear relation between the hydrogen content and the ductility of steel, as expressed by the percentage reduction in area. There are innumerable published examples of this in the literature. Hobson and Sykes[1] found that with ¼% C–3–3½% Cr–Mo low-alloy steels and also 1% Ni low-alloy steels, the inverse linear relation existed fairly exactly. The effect is quite large and for example, with the 3% Cr low-alloy steels, the percentage reduction in area fell from 74% to 34% for an increase in hydrogen content from 0 to 10 ml/100 g of steel. Hydrogen was found to have a similar effect upon the ductility of 0·7% Mn–0·1–0·4% Ni low-alloy steels but, as Fig. 5 shows, its effectiveness in reducing the ductility became less and less as the hydrogen content increased.

This tendency for hydrogen concentrations above a certain limit to have little further effect upon the ductility has been noticed by other investigators. Sims, Moore, and Williams[16] found that for 0·3% carbon steel the ductility fell to a nearly constant level at above a concentration of about 8 ml per 100 g of steel, thereafter decreasing at a very slow rate, as shown in Fig. 6. The idea of a limiting hydrogen content above which ductility remains constant at a low value (10–20%) has also found experimental support in the work of Seabrook, Grant, and Carney[13] on a 0·2% carbon steel. The elongation of this steel fell from 40% to 20% at 6½ ml/100 g, thereafter remaining constant at 20% up to hydrogen contents of 172 ml/100 g of steel.

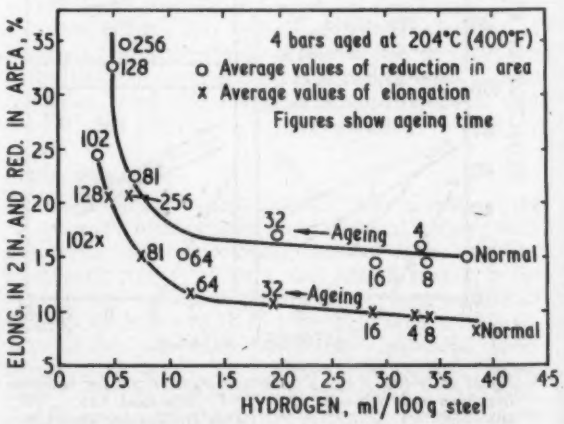
Other investigators have not specifically mentioned this effect, but in some of the published work hydrogen concentrations have not been measured specifically nor have they been as high as 8–10 ml/100 g. The curves in Fig. 3, after Hobson and Hewitt,[5] for 3% Cr–Mo low-alloy steel show some good examples of the inverse linear relation between hydrogen content and percentage reduction in area and, also, those at −78°C. show the other type of relation in which the ductility tends towards a nearly constant value. Herres[3] investigated five steels of varying carbon content and hardenability and found that hydrogen produced a marked fall in the percentage reduction in area; in some cases complete embrittlement occurred. Furnace-cooled specimens showed less embrittlement than quenched specimens, probably owing to almost complete loss of hydrogen during cooling.

In general, the effects of hydrogen in reducing the average elongation and reduction in area, as found by the previously mentioned workers, are qualitatively in agreement with the findings of Marshall, Garvey, and Llewellyn[4] on wrought AISI 1010 steel, of Vaughan and de Morton[18] on mild steel, of Shadburn, Marshall *et al.*[17] on 0·1% carbon steel, of Bastien and Anzou[6]

5—*Hydrogen embrittlement of two 0·35% C steels, showing effect of variation of max. stress and reduced effect of hydrogen as concentration rises. (After Hobson and Sykes[1])*



6—*Correlation of ductility to hydrogen content on ageing medium-carbon cast steel. (After Sims, Moore, and Williams[16])*



on carbon steels, and of Sims, Moore, and Williams[16] on Al-killed medium-carbon steel, Si-killed arc-furnace steel, and Si–Ti-killed, high-tensile basic open-hearth steel. Of the last three steels, the extent of reduction in ductility varied from steel to steel, and, indeed, this is true of all steels, but the high-tensile steel showed the steepest fall in ductility with hydrogen concentration. However, it is not general that, in a given group of steels, the fall in ductility is the greatest for the strongest steel, there being marked variability.

The austenitic steels constitute a special class which is much less embrittled by hydrogen than any other type of steel, and this may be attributable in some degree to the higher solubility of hydrogen in austenite. For example, Hobson and Sykes[1] found that even 10–12 ml of hydrogen per 100 g of steel reduced the ductility of 23% Cr–21% Ni and 12% Cr–12% Ni steels by only 4% and 15% respectively. In a more striking case, Hobson and Hewitt found that increasing the hydrogen content of 18% Cr–10%Ni steel from nearly zero to as high as 25 ml per 100 g of steel was virtually without effect upon the elongation and reduction in area at test temperatures between 20° and 400°C., except for a slight fall in the percentage reduction in area at 20°C. Under the same conditions, the percentage reduction in area fell from 66% to 46% at −78°C., although the elongation was only slightly reduced. However, Hobson and Hewitt attributed the embrittlement at −78°C. to the effect of ferrite produced by the breakdown of austenite under the cold work of the tensile test.

In general, the majority of ferritic steels recover their ductility on standing at room temperature for a period which shows that the effect of hydrogen is transient, and this would offer support for the theory that diffusible hydrogen (as opposed to residual) is the major factor in ductility. As examples of recovery of ductility, Hobson and Sykes[1] may again be cited. They found that on standing at room temperature for 21 days after charging with hydrogen, low-alloy steels recovered 85–90% of the ductility.

Although Herres[3] obtained qualitatively similar results on plain 0·25% carbon steels, which exhibited a recovery of ductility on ageing at room temperature for seven days, he found that a 0·4% carbon steel required tempering at 593°C. to achieve this degree of recovery, a result which is at variance with the results of other workers. Vaughan and de Morton[18, 19] obtained complete recovery of ductility in mild steel on ageing at room temperature.

Hence there seems to be some disagreement about the extent of the recovery of ductility on ageing at room temperature; some workers have obtained only partial recovery, and others complete recovery. In general, the thicker the section, the longer is the time required for diffusion of hydrogen out of the lattice and, for such large-diameter sections as steel rolls, room-temperatue diffusion of hydrogen is a lengthy process, often involving tens of years, so that some form of heat-treatment is required to eliminate hydrogen with any speed.

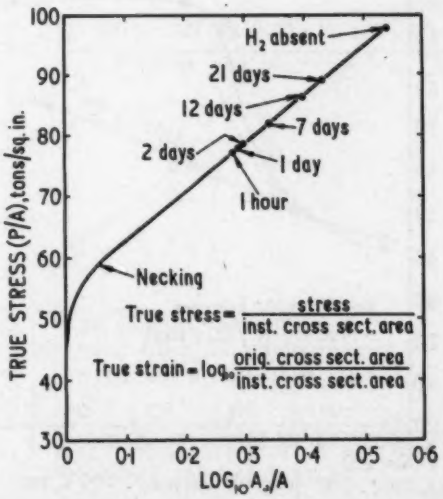
Apparent ultimate tensile strength, true breaking stress, yield point, and limit of proportionality

Whilst in a number of reported instances hydrogen appears to be without any effect on these three quantities, there are some in which hydrogen does appear to reduce apparent ultimate tensile strength slightly, and hydrogen is known to eliminate the yield point at room temperature and to reduce the proportionality limit under certain conditions.

An example of where hydrogen is without any effect is to be found in the results of Hobson and Sykes[1] already quoted. They found that with ¼% C–3/3½% Cr–Mo low-alloy steels, 0·7% Mn–0·1/0·4% Ni low-alloy steels, 12% Cr–12% Ni, and 23% Cr–21% Ni steels in a given heat-treated condition, variations in the hydrogen content were virtually without effect upon the yield point, proportionality limit, or apparent maximum stress. Similar results were found by Hobson and Hewitt[5] for ¼% C–3% Cr–Mo low-alloy steels and for 18% Cr–10% Ni steel.

The work of Herres[3] provides an example of an instance where the effect of charging hydrogen was to reduce slightly the apparent tensile strength, in this case, of five carbon steels of varying carbon content and hardenability. On the other hand, the yield point and the shape of the nominal stress/strain curve were uninfluenced by hydrogen, unless complete embrittlement occurred.

A number of investigators have established that hydrogen reduces the true breaking stress at fracture for quite a variety of steels, as for example, Shadburn and Marshall[17] for 0·1% carbon steel, Sims, Moore, and Williams[16] for 0·3% carbon steel, and Herres[3] and Bastein and Anzou[6] for a range of carbon steels. By using a MacGregor plot, in which true stress is plotted against true strain, as shown in Fig. 7, and in which these two quantities show a linear relation after the apparent maximum stress has been passed, Hobson and Sykes were able to show that hydrogen does not modify the true stress/strain curve until after the apparent maximum stress has been passed and necking is in progress. Herres[3] obtained a similar result. In Fig. 7, the diffusible hydrogen content is inversely proportional to the number of days of ageing, and it



7—True stress/strain curves for hydrogen-impregnated 3% Cr–Mo steel after standing in air for varying periods. (After Hobson and Sykes[1])

can be seen that diffusible hydrogen reduces the maximum stress in proportion to its own concentration. It seems likely that hydrogen acts by diffusing into the large number of slip planes and dislocation arrays available at this stage and reduces the total energy required for fracture.

Although it has been shown by Hobson and Sykes[1] that hydrogen does not affect the true stress/strain curve of low-alloy steels until necking has started, a number of investigators have shown that, with steels normally exhibiting an upper and lower yield point at room temperature, cathodically charging with hydrogen eliminates the yield point at room temperature and produces an increase in the slope of the initial part of the work-hardening portion of the apparent stress/strain curve. That hydrogen does eliminate the yield point has been shown by Cracknell and Petch[20] for iron and carbon steels up to 0·62% carbon, by Rogers[21] for Armco iron, rimmed steel, and partially for SAE 1020 steel, and by Vaughan and de Morton[18, 19] for mild steel. Rogers[22] has also shown that hydrogen develops its own yield point at below about −120°C., probably owing to anchoring of the normally mobile hydrogen atmospheres on dislocation sites, the hydrogen at these temperatures producing a yield point in the same way as nitrogen at higher temperatures.

In all the above-mentioned work[18, 19, 20, 21, 22] the hydrogen was charged cathodically, but it is to be regretted that the amount of hydrogen involved was not determined quantitatively, since it is not possible in consequence to estimate the initial amount of hydrogen which is required for elimination of the yield point.

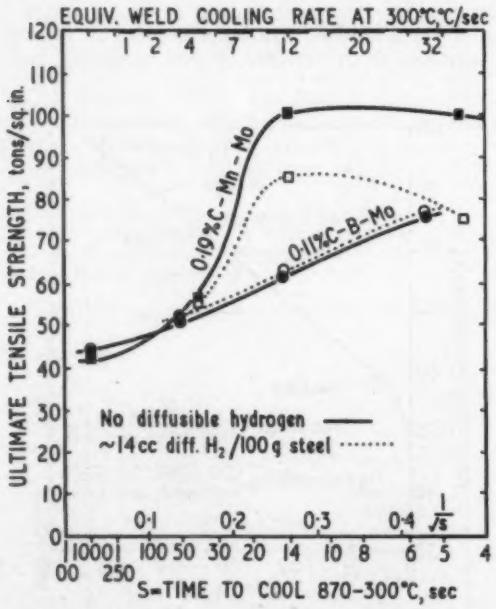
In contrast to other workers in this field, Vaughan and de Morton[18, 19] found that about 10 ml of cathodically charged hydrogen reduced the proportionality limit of mild steel from 13·8 to 12·7 tons/sq.in.; the reduction was also proportional to the hydrogen content. Hydrogen was found to produce a strain-ageing effect of its own; specimens strained 2% with and without hydrogen present, showed strain-ageing effects as defined by Leslie and Rickett[23] of 0·8% and 5% respectively. Moreover, in strain ageing, hydrogen caused a large reduction in the stress necessary for reappearance of the new yield point.

Method of introducing hydrogen

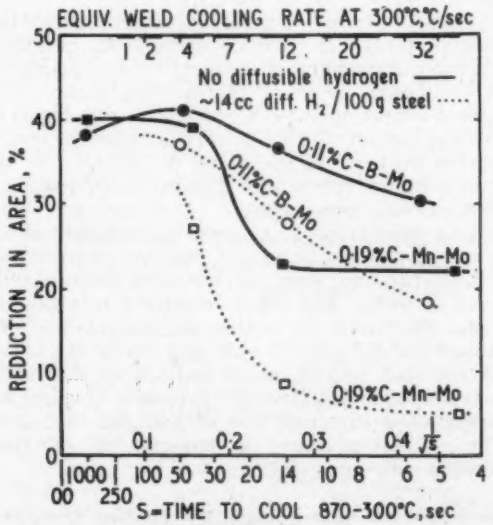
The work of Hobson and Sykes showed that hydrogen has the same embrittling effect upon low-alloy steels whether charged cathodically or by high-pressure methods; Herres[3] found that hydrogen charged cathodically or by soaking in a furnace at red heat at 1 atm pressure had the same embrittling effects upon the properties of five steels of varying carbon content and hardenability.

The work of Bastien and Anzou[8] also affords further proof that the manner of charging the hydrogen is immaterial with respect to the embrittling effect upon plain carbon steel. Their results show that hydrogen, charged either by electrolysis or by pickling in acid, produces identical embrittling effects. These results are qualitatively in agreement with the results of Herres,[3] but exact comparisons are impossible because hydrogen concentrations are not given.

It may be concluded that any differences due to the method of charging hydrogen are small compared with the inherent variability of hydrogen effects, as measured by present testing methods. By carrying out precise comparisons on samples of the same steel heat-treated in exactly the same way, charged with measured concentrations of hydrogen, and using a sufficient number of replicate samples for statistical analysis of the results, differences between the different charging methods may become detectable. For


EQUIV. WELD COOLING RATE AT 300°C,°C/sec
1 2 4 7 12 20 32
ULTIMATE TENSILE STRENGTH, tons/sq.in.
0·19%C-Mn-Mo
0·11%C-B-Mo
No diffusible hydrogen
~14cc diff. H2/100g steel
0·1 0·2 0·3 0·4 1/√s
1000 100 50 30 20 14 10 8 6 5 4
00 250
S-TIME TO COOL 870-300°C, sec

8—*Effect of variations in hydrogen content and cooling rate on U.T.S. of B–Mo and Mn–Mo steels. Plain specimens: peak temp. 1250–1350°C. (After Cottrell[15])*



9—*Effect of variations in hydrogen content and cooling rate on ductility of B–Mo and Mn–Mo steels. Plain specimens: peak temp. 1250–1350°C. (After Cottrell[15])*

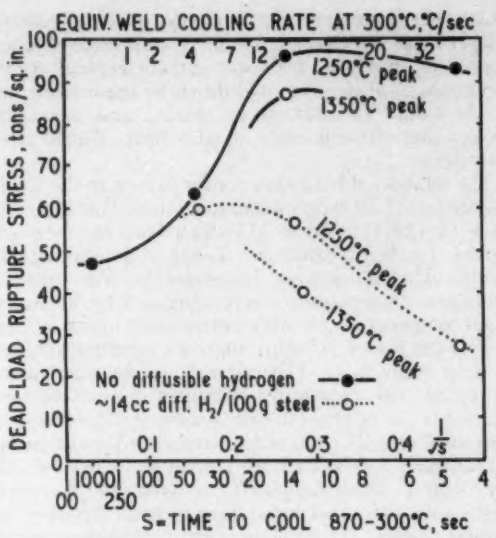
example, cathodic charging, even after a heat-treatment, may tend to produce a different type of hydrogen distribution from furnace charging of hydrogen at, say, 1100°C.

Effects on Tensile Properties of Weld Metal

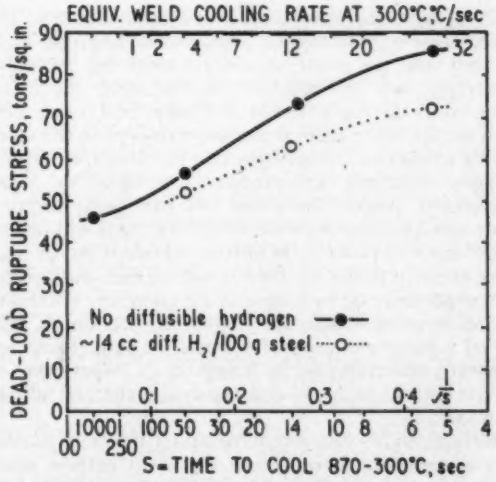
There are differences in the effects of hydrogen in steel and in weld metal and, in general, many of these differences may be attributed to the rapid heat-treatment cycle which weld metal undergoes during deposition. This is reflected in the microstructure of welds, which are know to differ from ordinary steels in their inherently fine grain size which is not affected by subsequent heat-treatments, and which may be the result of the large inclusion content of weld metal. Weld metal also has higher yield-point/ultimate-strength ratio than commercial steels of equivalent composition, a relatively high nominal ultimate strength, and a higher impact-energy value above the transition range than wrought steel. Hence it is not altogether surprising to find that hydrogen embrittlement of weld metal, although showing obvious similarities to that of ordinary steel, differs in many details and is generally treated quite separately.

For example, in contrast to steel, it can be seen from Cottrell's work[12] relating the tensile properties of simulated heat-affected zone metal to the weld cooling rate that hydrogen often leads to a deterioration of all the tensile properties. Three steels were tested at two levels of hydrogen concentration, namely zero and about 14 ml per 100 g as estimated from previous tests; the steels were cathodically charged and then subjected to a simulated welding cycle by means of high-frequency heating, weld cooling rates of between 1° and 32°C. per second being investigated. Although the effect of weld cooling rate upon the slow tensile properties is much greater than that of hydrogen, it is apparent from Figs. 8–12 that the embrittling effects of hydrogen are extended to the nominal ultimate tensile strength of the 0·19% C–Mn–Mo steel, in addition to its normal effect in reducing the ductility that occurs with both this steel and with the 0·11% C–B–Mo steel. This result may be reconciled with the results on wrought steels if it is remembered that these would fall on the slow-cooling section of the curve, where hydrogen is without effect upon the ultimate tensile strength. At normal weld cooling rates, the dead-load rupture stress was reduced by hydrogen for both of the above steels and also for a low C–Mn steel. In these tests by Cottrell there would be a rapid loss of hydrogen during the heating cycle.

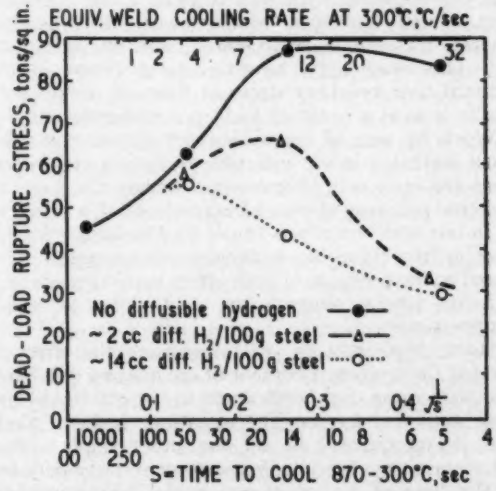
In all these examples where hydrogen had an embrittling effect, this increased proportionately with increase in the weld cooling rate. The concentrations of hydrogen used in these tests were approximately those


EQUIV. WELD COOLING RATE AT 300°C, °C/sec
1 2 4 7 12 20 32
1250°C peak
1350°C peak
1250°C peak
1350°C peak
DEAD-LOAD RUPTURE STRESS, tons/sq. in.
No diffusible hydrogen
~14 cc diff. H2/100 g steel
0·1 0·2 0·3 0·4 1/√s
1000 100 50 30 20 14 10 8 6 5 4
00 250
S=TIME TO COOL 870-300°C, sec

11—*Effect of variations in hydrogen content and cooling rate on dead-load rupture stress of Mn–Mo steel. Notched specimens. (After Cottrell[13])*


EQUIV. WELD COOLING RATE AT 300°C, °C/sec
1 2 4 7 12 20 32
DEAD-LOAD RUPTURE STRESS, tons/sq. in.
No diffusible hydrogen
~14 cc diff. H2/100 g steel
0·1 0·2 0·3 0·4 1/√s
1000 100 50 30 20 14 10 8 6 5 4
00 250
S=TIME TO COOL 870-300°C, sec

10—*Effect of variations in hydrogen content and cooling rate on dead-load rupture stress of low-carbon B–Mo steel. Notched specimens: peak temp. 1350°C. (After Cottrell[13])*


EQUIV. WELD COOLING RATE AT 300°C, °C/sec
1 2 4 7 12 20 32
DEAD-LOAD RUPTURE STRESS, tons/sq. in.
No diffusible hydrogen
~2 cc diff. H2/100 g steel
~14 cc diff. H2/100 g steel
0·1 0·2 0·3 0·4 1/√s
1000 100 50 30 20 14 10 8 6 5 4
00 250
S=TIME TO COOL 870-300°C, sec

12—*Effect of variations in hydrogen content and cooling rate on dead-load rupture stress of C–Mn steel to BS.968a. Notched specimens: peak temp. 1300°C. (After Cottrell[13])*

which are found in the deposits of electrodes to BS.1719 (1951) Classes 2 and 3 and some Class 1 electrodes, but the steels do not correspond to the composition of deposits laid down by these electrodes which would be difficult to match, and hence the results are relevant only to the heat-affected zone material.

The relation of hydrogen concentration to the tensile properties of all weld-metal specimens laid down by BS.1719 (1951) class E.217 electrodes has been reported by Winterton[26] in a paper presenting the results of a co-operative investigation. Variations in hydrogen concentration were obtained by ageing at room temperature for different periods up to 76 days and by the use of different interpass temperatures and welding techniques. The ductility of the weld metal decreased with increasing hydrogen content but not uniformly, as is general with ordinary steels. In contrast to Cottrell's results[12] on simulated weld metal, the ultimate tensile strength tended to increase as the hydrogen content increased, as also did the yield point, this last effect never having been reported for ordinary steels. The non-linearity of all these relations is attributable to welding variables. Similar variable results were obtained with deposits from BS.1719 Classes E.110, E.317, E.435 and E.537 electrodes but, in some cases, the previously unreported result was obtained, in that the ductility increased with increasing hydrogen concentration.

Further conflicting evidence is provided by the work of Flanigan,[25] who investigated the effect of hydrogen on mild-steel arc-weld deposits by means of a notch-bend test at −40°C. In regard to the hydrogen introduced into welds laid down by E.6010 and E.6015 electrodes, his evidence for a relation between hydrogen content and ductility was inconclusive. Hydrogen charged by pickling into hydrogen-free weld metal laid down by E.6010 electrodes did have an embrittling effect, the bend angle being reduced by 40–50% for only 2·3 ml hydrogen per 100 g. The ductility recovered fully on post-heating pickled welds for 3 hr at +343°C. The implication of this work is that the diffusible hydrogen content is the operative part of the hydrogen which lowers the ductility, whereas the residual hydrogen is relatively unimportant. The latter may, however, be a factor in the reduced true breaking stress at fracture obtained in tensile tests as a result of hydrogen embrittlement.

Work by one of the authors,[26a] shows that even small increases in the diffusible hydrogen concentrations are extremely effective in lowering the ductility and true breaking stress at fracture, and this seems to be in line with the effects found by Flanigan.[25] Variations in the diffusible hydrogen concentration were found to have relatively little effect upon the nominal ultimate tensile strength or yield point at room temperature.

Stout, McGeady, *et al.*[2] investigated the effect of varying the hydrogen content of the coating of a basic electrode upon the bend angle in longitudinal-bend slow notch-bend tests. The variations in bend angle were insignificant but, on the other hand, direct determinations of the hydrogen content of the deposits at the time of test were not made. These workers attempted to assess the effects of hydrogen by means of a comparison between the high-hydrogen type E.6010 deposits and the low-hydrogen type basic electrode deposits, but this is not a valid comparison in view of the other differences between these electrodes in such important factors as oxygen content, manganese content, and the like. The important conclusion is reached that hydrogen embrittlement is of a temporary nature in many weldments, which offers support for the view that diffusibly hydrogen is the most important factor in hydrogen embrittlement. The corollary to this is that hydrogen embrittlement of steels in service is considerably less than during and shortly after welding fabrication.

The effect of 10 ml per 100 g of cathodically charged hydrogen upon the mechanical properties of BS.1719 Class 3 and Class 6 electrodes deposits was investigated by Vaughan and de Morton.[18] The reduction in area fell from 62 to 15% in both cases, the U.T.S. showed a fall of 7% in the Class 3 deposit but none in the Class 6 deposit, and the limit of proportionality was raised by about 1 ton/sq.in. Since the Class 3 deposit contained about 10 ml/100 g of residual hydrogen in both cases, the effect of this upon the mechanical properties cannot be predicted with certainty.

Effect on Fatigue Strength of Steel

Because of the experimental difficulties, little study has been made of the effect of hydrogen on the fatigue strength of steel.

In fatigue tests upon 0·3% Cr–3% Cr–Mo low-alloy steel, in which the stress was adjusted to cause failure within 20 min, thus minimising the loss of hydrogen, increasing the hydrogen content from 0·8 to 7 ml per 100 g, has been found to reduce the number of reversals by about two-thirds.[5]

Effect on Cracking of Steels and of Ferritic and Stainless-steel Welds and Other Related Defects due to Hydrogen

Wild ingots, blisters, porosity, pin-holes, and seams in steel ingots

Hydrogen has been established as a factor in the production of wild ingots. These occur when the steel poured into an ingot mould contains an excess of hydrogen over the solubility in the steel; the metal rises up in the ingot mould and may 'boil over'. This also occurs when there is excessive oxygen in the steel, owing to the evolution of gas as a result of the carbon-oxygen reaction, or excessive nitrogen in high-chromium steels. There are, in fact, also limiting hydrogen contents between which an ingot will not be 'wild' but will suffer from porosity. Porous ingots may even seem to freeze in the normal manner but, owing to the presence of hydrogen in excess of the solubility in the freezing range, the ingots, on sectioning, will reveal a pattern of pin-holes under the skin. These are generally accompanied by a number of larger holes in the centre of the ingot and traversing the crystallization axes.

Barraclough[27] has given the upper limits of hydrogen concentration for sound ingots in carbon steel, low-alloy steel, chromium stainless steel, and austenitic stainless steel as 6·5, 7·0, 10·0 and 12·0 ml/100 g respectively, and the ranges of hydrogen content for

porous ingots are 6·5–8·5, 7–9, 10–12 and 12–14 respectively, hydrogen contents above the upper limiting value leading to wild ingots. It is well known that the gas liberated during solidification does in fact consist of carbon monoxide, nitrogen, and hydrogen and, if these gases are present in significant quantities, the stated limits are of little value. If nitrogen is present, the hydrogen limits are proportionately lower and, if the steel is unkilled, it is difficult, if not impossible, to give any limiting value for the hydrogen content in practice.

Where ingot moulds have been 'tarred' with insufficient care, as when the mould is cold and too much tar is applied, or if the tar is moist, local evolution of hydrogen from the tar may result in a series of pinholes just beneath the surface of the freezing ingot. On rolling the ingot to billet, the same effects, but on a smaller scale, are obtained as in porous ingots—the billet is covered with a series of fine seams more or less parallel and relatively shallow, which are liable to open up if not removed before forging.

The general cure for all these related defects is to keep the hydrogen level at a minimum by drying all materials and components which come into contact with the liquid steel. There is considerable literature on the subject of this section, and the details have been investigated and described by a number of workers.[27, 28, 29, 30, 31, 32]

Transformation, hairline cracking, and weld fissuring

Cottrell[33] has shown that about 14 ml of hydrogen per 100 g of steel can reduce the end-of-transformation temperature (M_f) in cooling simulated weld specimens of a 0·19% C–Mn–Mo steel from about 240° to about 180°C. for cooling rates between about 7° and 25°C. per second at 300°C. Cottrell[34] stated that except at the highest cooling rates of about 32°C./sec, when the end-of-transformation temperature was reduced from 300° to 220°C., hydrogen was without effect in reducing the end-of-transformation temperature of a 0·11% C–B–Mo steel. On the other hand, Vaughan and de Morton[18] found that 8–12 ml of hydrogen per 100 g of weld metal had no effect upon the end-of-transformation temperature of a mild-steel weld deposit.

Andrew, Lee, Mallik, and Quarrell[35b] found that for a 3% Ni–Cr–Mo steel austenitized at 1150°C. for four days, hydrogen retarded transformation under isothermal conditions in the pearlite range (600°–450°C.) but had no effect in the intermediate range and slightly accelerated transformation in the lower bainite range by reducing the incubation period.

Kumar and Quarrell[36] studied the effect of hydrogen at the lower temperatures of the austenite to martensite transformation, *i.e.*, below 350°C. For concentrations in the range 4–7 ml/100 g, hydrogen was found to:

(*a*) Lower the end of transformation temperature by about 20°–40°C. of 0·3% C–3% Ni–Cr steel, 0·29% C– 4½% Ni–Cr–Mo steels, and certain C–Cr steels

(*b*) Have no effect on 0·2/0·3% C–3% Cr–Mo steels or 0·55/0·8% C–4½% Ni–Cr–Mo steels

(*c*) Lower the M_s temperature by as much as 50°C. and simultaneously raise the M_f temperature for 0·57–0·86% C–3% Cr–Mo steels, the delayed transformation occurring with a pronounced 'burst'.

Kumar and Quarrell[36] also found that hydrogen accelerated the isothermal transformation of 0·57% C–3% Cr–Mo steels in the lower bainite range at 273°C., the normal incubation period of 9 min being reduced to nil.

The effect of hydrogen upon the transformation behaviour of steel increases with increasing carbon content, which can be explained tentatively as a tendency to prevent 'carbon clustering' in higher-carbon steels; this drastically reduces the amount of carbon-free regions of austenite available which, according to the nucleation and growth theory, act as embryos for martensite formation, leading to undercooling below the normal M_s temperature. Hence when transformation does occur, it proceeds in a pronounced 'burst'.

Hairline crack formation was studied in a series of papers[35, 36] by Andrew and Quarrell. In 1946 Andrew, Quarrell, *et al.*[35a] found that it was always associated with hydrogen, more particularly with a hydrogen-rich constituent, and that transformation stresses were important in determining the crack distribution; they also found some cases of hairline cracking when transformation stresses were absent. However, in 1955 Troiano, Dana, and Shortsleeve[37] concluded that stresses due to low-temperature transformations were essential to the hairline cracking phenomena. Both Troiano *et al.* and Andrew *et al.* used steels which were hydrogenized at austenitizing temperatures, followed by isothermal tempering in a hydrogen atmosphere and quenching at room temperature.

On the other hand, Barker and Wainwright[38] charged hydrogen into a number of steels at 100 atm pressure at sub-critical temperatures, namely 600°–650°C., followed by air cooling, so that the austenite/martensite transformation stresses were not present, and residual stresses were low. In spite of this, 3/4% Ni–Cr–Mo steels showed hairline cracking but not plain 0·1% carbon steels. Similar results were obtained by Houdremont and Schrader,[39] which suggests that hydrogen is the most important factor in hairline cracking.

Kumar and Quarrell[36] pointed out that where the delayed austenite/martensite transformation occurs with a 'burst', the low-temperature stresses accompanying the latter may well be sufficient to trigger off hairline cracks, when superimposed on the pre-existing transformation and cooling stresses. These stresses, coupled with pressure due to hydrogen in the voids of the steel and that precipitated from supersaturated martensite, together with the embrittling effect of hydrogen on martensite, would complete the process of hairline cracking.

The rather similar phenomenon of weld fissuring has been shown by a number of investigators[18, 40, 41, 42, 43] to involve the presence of hydrogen, but the explanation of this phenomenon depends on whether austenite can be retained in low-carbon steels down to 120°C., the initial temperature of weld fissuring according to Flanigan.[43] That this may happen is suggested by the results of Averbach, Castleman, and Cohen,[44] who showed that 0·2% carbon steels can retain austenite down to room temperature. From their work it appears that when low-carbon steel is austenitized at high temperatures and drastically quenched, as in welding, some austenite will be retained. This tendency

would be accentuated if carbon segregation occurs in weld metal, such as was found by Thomason[45] in weld boundary zones. If this is so, even 0·1% carbon weld metal may have high-carbon islands in which traces of austenite are retained down to 120°C. In addition, weld fissuring may occur in a similar manner to hair-line cracking, as already described, with the added factor that larger amounts of hydrogen may be present in both the retained austenite and the voids, and also more drastic cooling and transformation stresses may be present, owing to the quench effect which occurs in welded joints.

Cold cracking in welded joints

Cold cracking, or under-bead cracking as it is often called, is known to occur at room temperatures in the heat-affected zones of welds in the hardenable steels, under certain adverse conditions. Work by a number of investigators[3, 41, 42, 46, 47, 48, 49, 50] has shown that hydrogen is one of the major factors in promoting cracking of this nature. But although there is general agreement that other major factors are the harden-ability of the base metal (which involves the carbon and alloy metal contents), the degree of restraint of the welded joint, the base plate thickness, and the cooling rate of the weld, there is still disagreement as to the precise mechanism of cold cracking. It is known that the lower the hydrogen content of the weld metal deposited by an electrode, the less is the risk of cold cracking.

Hot cracking of welds

There is conflicting evidence on whether hydrogen is a factor in the hot cracking of weld metal. Apold[51] was unable to find any relation between the length of hot cracks in welds deposited by ten basic electrodes and the diffusible hydrogen concentrations, which ranged from 2·0 to 10·3 ml per 100 g. It is impossible to decide whether or not hydrogen is a possible second-order factor in the cracking of basic electrode deposits on the basis of these results because no information was presented on the Mn/S ratio of the deposits or the amounts of carbon, silicon, and nickel present, all of which are suspected to be factors in hot cracking.

The metallurgical factors involved in hot cracking have been investigated by Rollason and Roberts,[52, 53] for ferritic steel welds, who were able to eliminate hot cracking completely by ensuring very high Mn/S ratios (over 30:1), low nickel contents, and careful control of the elements carbon, silicon, and phosphorus. Hydrogen was not investigated specifically, so that here again it is not possible to decide whether it could be a second-order effect or not, and this also applies to stainless steel weld metals, which were investigated by Rollason and Bystram.[54]

Other workers in this field have produced evidence in favour of hydrogen as a factor. Winterton and Cottrell[55] found a rough correlation between the hot-cracking tendency in the BWRA double-fillet test and the hydrogen content of the weld metal, although it is possible that the Mn/S ratio and other factors were also subject to variations in these tests. In contrast to Apold,[51] who found that baking basic electrodes before welding was without effect upon hot cracking, Winterton and Cottrell[55] reported reduced hot cracking on baking the electrodes and increased cracking on humidifying the electrode coating. It has been claimed that basic electrodes of low potential hydrogen content are less susceptible to hot cracking than the higher hydrogen-bearing rutile and oxide silicate types, but this superiority of the basic type also coincides with much higher Mn/S ratios in their deposits. It has also been found that higher welding currents for the same electrode, which lead to higher hydrogen concentrations in the weld deposit, are accompanied by more hot cracking.

Formation of haloes and snowflakes in tensile specimens

Haloes are small oval or circular bright areas containing roughly spherical cavities and often exhibiting radial markings from the nuclei; they are found on the fractured surfaces of weld metal tensile specimens; fissures and elongated cavities running parallel to the fractured surface often occur.

Winterton[56] has summarized most of the existing information on halo formation in weld metal tensile specimens. There is more or less general agreement on the fact that halo formation occurs only in the presence of a certain minimum quantity of hydrogen, of straining conditions, as in the normal tensile tests, which are slow enough for the movement of hydrogen to occur through the strained lattice, and of a nucleus such as a gas pore or inclusion; inclusions of the rhodonite type, such as are encountered with rutile electrodes, are commonly but not necessarily associated with halo formation. The fact that movement of hydrogen is involved in the production of haloes seems to be established by the fact that haloes are not encountered in tensile specimens which have been tested at very high strain rates or temperatures below about −120°C. or annealed *in vacuo* at 650°C. These are the conditions which, in the first case, do not allow time for hydrogen diffusion, in the second case, immobilize the hydrogen, and in the third case remove all the hydrogen.

Although a number of 'theories' of halo formation have been advanced, it would be more correct to describe them as hypotheses, and development of a sound theory will involve more detailed work on the precise straining conditions, the functions of voids and inclusions, and the concentrations of hydrogen involved, together with information on the concentration gradients and the type and rate of 'movement' of hydrogen required by such a theory.

A closely related phenomenon is involved in the production of snowflakes which are encountered in the tensile testing of steels containing hydrogen in cavities. On fracture of the test piece, concentric brittle zones are found around gas pockets, the remainder of the fracture being ductile. Zapffe[57] has described many typical examples of haloes or fish-eyes and snowflakes, and has discussed the origin of these defects in some detail.

Theories of hydrogen embrittlement

Although a large number of theories have been presented over the past decade[7, 57–64] it seems likely that no comprehensive and satisfactory theory of hydrogen embrittlement can be developed in the present state of knowledge on the subject.

In a paper on the subject by Morlet, Johnson, and Troiano,[64] they suggest that voids in a steel are

potential stress raisers, so that application of a stress creates a multi-axial stress state in the lattice volume around each void. There is a point of maximum triaxiality, at which the stress state will be of the greatest severity. They then postulate that the hydrogen concentration in this region of greatest triaxiality controls the degree of embrittlement. These writers do not, however, claim to have obtained a complete theory of hydrogen embrittlement.

Hardness

Until recently it has been generally accepted that hydrogen is without effect upon the hardness of steel.[13, 24, 65, 66] This has been demonstrated by Seabrook, Grant, and Carney[13] for 0·2% carbon steels; according to Zapffe and Haslam,[24] hydrogen charged cathodically into 17% Cr–1% C stainless-steel wire had no effect upon the hardness. Hydrogen charged by pickling in 10% H_2SO_4 at 49°C. was shown by Flanigan[25] to have no effect upon the hardness of welds laid down by E.6010 electrodes, but the amount of hydrogen introduced was only 2·33 ml/100 g.[25] In 1942 Chaudron and Moreau[67] reported a slight increase in hardness as a result of electrolytic charging of hydrogen into steel, which was not obtained on charging by pickling only. Recently, however, Vaughan and de Morton[18] using a long indentation time of 60 sec in a Kentron hardness machine, observed a reversible hardening effect in mild steel (0·17% C, 0·5% Mn) due to hydrogen. As the specimen aged at room temperature, the loss of 3 ml of diffusible hydrogen per 100 g was reflected in a fall in hardness from 130 to about 114 Brinell after 5 days, the latter value being the hardness of the original steel. This work suggests that, as with tensile testing, there is a time factor involved in the effect of hydrogen on hardness, which may be related in some way to the time for diffusion of hydrogen to occur to points of triaxial stress concentration, dislocation arrays, and so on.

Summary and Conclusions

It is clear that hydrogen causes a deterioration in the impact properties of some steels but not of others, although at speeds of impact testing above about 10,000 in./in./min, embrittlement does not occur. Whether embrittlement occurs or not depends upon the hydrogen concentration, the microstructure of the steel, the test temperature, the type of test, the speed of the test, and the alloying elements present in the steel. The relation between hydrogen embrittlement and variation in the microstructure of a given steel is perhaps the most important variable still requiring elucidation. Similar remarks apply to the impact properties of weld metal, but here in particular more quantitative determinations are needed of the diffusible atomic hydrogen present in solution in the lattice, and that present in molecular form in voids and discontinuities (residual hydrogen) in relation to variations in other properties. Much of the existing work is based on an assumed hydrogen concentration; yet, in the absence of information on the concentrations of the two types of hydrogen, it will be impossible to develop a complete theory of hydrogen embrittlement.

The tensile properties of steels, as affected by variations in the hydrogen concentrations present, have received considerable attention but, again, without much attempt to discriminate between the effects of diffusible atomic and residual hydrogen. It is probable that many of the anomalies reported, concerning the effects of hydrogen, would be removed if the degree of embrittlement were related not to the total hydrogen concentration, but to the concentration of one or other particular form of hydrogen.

In ferritic steels, hydrogen reduces the ductility and the true breaking stress at fracture, but the embrittlement varies considerably with microstructure, testing speed, temperature, and alloying elements present. This is also largely true of weld metal, with some exceptions, although weld metal exhibits an inherently greater degree of variability than wrought steel. In many instances, hydrogen is without much affect upon the nominal tensile strength and yield point or proportionality limit of wrought steels, although there are some reported cases in which the proportionality limit is reduced. For weld metal there is conflicting evidence for both a strengthening and an embrittling effect of hydrogen upon the tensile properties, which is possibly a reflection of the larger variations encountered in the microstructure of welds. A point of particular interest is that, in some instances, the high cooling rates of weld metal are reported by some workers to be reflected in a stengthening effect of hydrogen upon the nominal ultimate tensile strength.

The elimination of the yield points at room temperature by hydrogen, and the production of a hydrogen yield point below −120°C. in certain types of carbon steels requires further investigation, so as to find the correct explanation of the phenomena; it appears that hydrogen can have a strain-ageing effect also. One factor in particular, which still requires to be investigated, is the minimum concentration of hydrogen required for the occurrence of these yield-point phenomena in wrought carbon steels. This is a point of some interest. It appears that the diffusible hydrogen normally present in steel or weld metal does not lead to suppression of the yield point, possibly because it differs from cathodically charged hydrogen, in that it is charged initially into the liquid metal, so that plastic deformation of the metal cannot occur during charging.

It is considered that future research into the effects of microstructure upon hydrogen embrittlement would be more rewarding perhaps than detailed studies of defects such as haloes in weld metal, and porosity in ingots, concerning which much information already exists. Further work on the effect of hydrogen on the transformation characteristics of steel and weld metal is required.

Acknowledgments

This paper is published by kind permission of the Directors of Murex Welding Processes Limited.

REFERENCES

1. J. D. Hobson and C. Sykes: *J. Iron Steel Inst.*, 1951 vol. 169, p. 209.
2. R. D. Stout, L. J. McGeady, C. P. Sun, J. F. Libsch, and G. E. Doan: *Welding J.*, 1947, vol. 26, p. 335s.

3. S. A. Herres: *Trans. Amer. Soc. Metals*, 1947, vol. 39, p. 162
4. S. Marshall, T. M. Garvey and D. S. Llewelyn: *Proceedings Electric Furnace Conf. A.I.M.E.*, 1948, vol. 6, p. 59.
5. J. D. Hobson and J. Hewitt: *J. Iron Steel Inst.*, 1953, vol. 173, p. 130.
6. P. Bastien and P. Anzou: *Comptes Rendues*, 1949, vol. 229, p. 549.
7. J. T. Brown and W. M. Baldwin: *J. Met.*, 1954, vol. 6, p. 298.
8. W. P. Rees and B. E. Hopkins: *J. Iron Steel Inst.*, 1952, vol. 172, p. 403.
9. W. P. Rees, B. E. Hopkins and H. R. Tipler: *J. Iron Steel Inst.*, 1951, vol. 169, p. 157.
10. B. E. Hopkins and H. R. Tipler: *J. Iron Steel Inst.*, 1954, vol. 177, p. 110.
11. W. Pellini and E. W. Eschbacher: *Welding J.*, 1954, vol. 33, p. 16s.
12. C. L. M. Cottrell: *Brit. Welding J.*, 1954, vol. 1, p. 167.
13. J. Seabrook, N. J. Grant and D. Carney: *Trans. Amer. Inst. Min. Met. Eng.*, 1950, vol. 188, p. 1317.
14. N. J. Petch and P. Stables: *Nature*, 1952, vol. 169, p. 842.
15. P. Bastien and P. Anzou: *Comptes Rendues*, 1951, vol. 232, p. 69.
16. C. E. Sims, G. A. Moore and D. W. Williams: *Metals Tech.*, 1948, vol. 15, T.P. 2454; *Trans. Amer. Inst. Min. Met. Eng.*, 1948, vol. 176, p. 283.
17. R. Shadburn, S. Marshall, *et al.*: *Proceedings Electric Furnace Conf. A.I.M.E.*, 1948.
18. H. G. Vaughan and M. E. de Morton: *Brit. Welding J.*, 1957, vol. 4, p. 40.
19. H. G. Vaughan and M. E. de Morton: *Acta Met.*, 1956, vol. 4, p. 224.
20. A. Cracknell and N. J. Petch: *Acta Met.*, 1955, vol. 3, p. 200.
21. H. C. Rogers: *Acta Met.*, 1956, vol. 4, p. 114.
22. H. C. Rogers: *Acta Met.*, 1954, vol. 2, p. 167.
23. W. C. Leslie and R. L. Rickett: *J. Met.*, 1953, vol. 5, p. 1021.
24. C. E. Zapffe and M. A. Haslam: *Metals Tech.*, 1946, vol. 13.
25. A. E. Flanigan: *Welding J.*, 1947, vol. 26, p. 193s.
26. K. Winterton and E. G. P. Hinds: *Brit. Welding J.*, 1955, vol. 2, p. 513.

26a. P. D. Blake: *Brit. Welding J.*, 1958, vol. 5, pp. 565-567.

27. K. C. Barraclough: *Murex Review*, 1954, vol. 1, p. 305.
28. C. R. Garr and A. R. Troiano: *Trans. Amer. Inst. Min, Met. Eng.*, 1957, vol. 209, p. 445.
29. C. A. Zapffe and C. E. Sims: *Metals and Alloys*, 1939, vol. 11, p. 145.
30. E. Houdremont and H. Korscham: *Stahl u Eisen*, 1935, vol. 55, p. 297.
31. H. Benneck, H. Schenck and H. Muller: *Stahl u Eisen*, 1935, vol. 53, p. 321.
32. C. E. Sims: *Proceedings Electric Furnace Conf. A.I.M.E.*, 1949, vol. 7, p. 302.
33. C. L. M. Cottrell: *Brit. Welding J.*, 1954, vol. 1, p. 167.
34. C. L. M. Cottrell: *Brit. Welding J.*, 1954, vol. 1, p. 315.

35a. J. H. Andrew, H. Lee, A. K. Mallik and A. G. Quarrell: *J. Iron Steel Inst.*, 1946, vol. 153, p. 67.

35b. J. H. Andrew, H. Lee and A. G. Quarrell: *J. Iron Steel Inst.*, 1942, vol. 146, p. 193.

36. R. Kumar and A. G. Quarrel: *J. Iron Steel Inst.*, 1957, vol. 187, p. 195.
37. A. R. Troiano, A. W. Dana and F. J. Shortsleeve: *Trans. Amer. Inst. Min. Met. Eng.*, 1955, vol. 203, p. 895.
38. A. Barker and A. Wainwright: *Nature*, 1956, vol. 177, p. 1136.
39. E. Houdremont and H. Schrader: *Stahl u Eisen*, 1941, vol. 27, p. 653.
40. E. C. Rollason: *J. Iron Steel Inst*, 1950, vol. 166, p. 105.
41. P. D. Blake: *Brit. Welding J.*, 1957, vol. 4, p. 146.
42. P. D. Blake: *The Welder*, 1955, vol. 24, p. 14.
43. A. E. Flanigan and T. Mideu: *Welding J.*, 1953, vol. 32, p. 99s.
44. B. L. Averbach, L. S. Castleman and M. Cohen: *Trans. Amer. Soc. Met.*, 1950, vol. 42, p. 112.
45. H. Thomason: *Proceedings British Commonwealth Welding Conference*, 1957, p. 275.
46. S. L. Hoyt, C. E. Sims and H. M. Banta: *Trans. Amer. Inst. Min. Met. Eng.*, 1945, vol. 162, p. 326.
47. M. W. Mallett and P. J. Rieppel: *Welding J.*, 1946, vol. 25, p. 396s.
48. E. C. Rollason: *Trans. Inst. Welding*, 1944, vol. 7, p. 74.
49. G. L. Hopkins: *Trans. Inst. Welding*, 1944, vol. 7, p. 76.
50. P. D. Blake: *The Welder*, 1956, vol. 25, p. 35.
51. A. Apold: *Welding Research*, 1952, vol. 6, p. 58r.
52. E. C. Rollason: *The Welder*, 1953, vol. 22, p. 45.
53. E. C. Rollason and D. F. T. R. Roberts: *Welding Research*, 1950, vol. 4, p. 129r.
54. E. C. Rollason and M. C. T. Bystram: *J. Iron Steel Inst.*, 1951, vol. 169, p. 347.
55. K. Winterton and C. L. M. Cottrell: *Metallurgia*, 1954, vol. 49, p. 3.
56. K. Winterton: *Brit. Welding J.*, 1955, vol. 2, p. 385.
57. C. A. Zapffe and C. E. Sims: *Trans. Amer. Inst. Min. Met. Eng.*, 1941, vol. 149, p. 225.
58. R. P. Frohmberg, W. J. Barnett and A. R. Troiano: *Trans. Amer. Soc. Metals*, 1955, vol. 47, p. 892.
59. N. J. Petch: *Phil. Mag.*, 1956, vol. 1, p. 331.
60. N. J. Petch and P. Stables: *Nature*, 1952, vol. 169, p. 842.
61. C. Zapffe: *Trans. Amer. Soc. Metals*, 1947, vol. 39, p. 191.
62. P. Bastien and P. Anzou: *Proc. 1st World Metal Congress*, 1951, pp. 535–552.
63. F. de Kazinczy: *J. Iron Steel Inst.*, 1954, vol. 177, p. 85.
64. W. J. Barnett and A. R. Troiano: *J. Met.*, 1957, vol. 209, p. 486.

64a. J. G. Morlet, H. H. Johnson and A. R. Troiano: *J. Iron Steel Inst.*, 1958, vol. 189, p. 37.

65. P. Mabb: *Metallurgia*, 1937, vol. 15/16, p. 141.
66. G. Chaudron, A. Portevin and L. Moreau: *Comptes Rendues*, 1938, vol. 218, pp. 207–235.
67. G. Chaudron and L. Moreau: *Korrosion u Metallschutz*, 1942, vol. 18, 134.

Welding
RESEARCH

B.W.R.A. REPORT

Metallurgical Examination of Oxygen-Cut Titanium

By J. C. Borland, A.I.M., and W. G. Hull, F.I.M.

Oxygen-cut titanium has been metallurgically examined, and various zones of contamination have been detected. Hardness surveys showed that the most severe hardening, and thus contamination, was restricted to the flame-cut surface. Titanium sheet 0·2 in. thick was flame-cut and butt welded along the line of the cut. The contamination of the weld metal was small, and the increase in hardness was only 38 D.P.N.

RECENT work by Coates[1] has shown that titanium may be cut by the oxygen-cutting process and, having once started, the cut can be made to progress at least five times as fast as in steel of similar thickness. Machining is more laborious than flame cutting, and whereas a 9 in. dia. billet may take hours to saw, flame cutting would only take a few minutes.

In view of the possibility of the oxy-cutting method becoming worthwhile, particularly where suitable machine tools are not available, it is desirable to have a knowledge of the metallurgical problems associated with cutting, and in particular to what extent oxygen contamination of the material occurs. Since it has been suggested[1] that flame cutting might in addition be used for preparing edges for welding, a few experiments were also carried out to assess the extent of contamination caused by the scale dissolving in the weld bead.

Materials

The titanium samples flame-cut and examined comprised a 9 in. dia. billet of specification Ti–150*A* (2·7% Cr, 1·3% Fe), a 3 in. square billet of *A*110–*AT* material (5% Al, 2½% Sn), and two lengths of unalloyed titanium, 0·20 in. and 0·125 in. thick.

The Ti–150*A* material is an $\alpha\beta$ type alloy, which hardens appreciably on rapid cooling. The two other materials belong to the α group, and are not expected to exhibit hardening characteristics in the same sense but, as is pointed out later, some hardening at high cooling rates does in fact occur.

The cutting speed was 20 in./min for the 9 in. dia. billet, and 96 in./min for the titanium sheets. The cutting speed for the titanium–aluminium–tin alloy was unknown. The actual conditions for making these cuts do not necessarily represent the optimum conditions which can be achieved, and so further details of the process are omitted here.

Gas analysis

An analysis of the commercially pure titanium was made after cutting, to determine the overall contamination and also to confirm that oxygen was the principal gas absorbed. The results which are summarized in Table I, refer to the parent sheet, and to a

Report A.4/2/58 of the B.W.R.A., issued to members in August, 1958.

Mr. Borland is Head of the Reactive Metals Group at the British Welding Research Association. Mr. Hull, formerly with the Association, is now Head of the Metallurgical Laboratories of Atomic Power Constructions Ltd.

Table I

Gas analysis of unalloyed titanium flame-cut samples

Thickness, in.	*Sample*	*Gas Contents* O_2, %	N_2, *p.p.m.*	H_2, *p.p.m.*
0·2	A.1, Parent sheet	0·067	240	55
	A.1, Contaminated slice	0·32	230	60
0·125	A.2, Parent sheet	0·11	250	55
	A.2, Contaminated slice	0·30	230	65

slice 0·030 in. thick taken from the contaminated zone so as to include scale, core, and unaffected material.

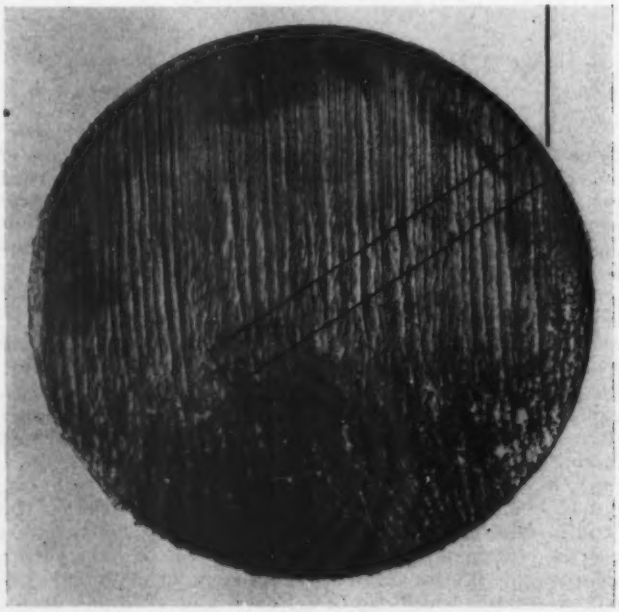
It is seen that the bulk of the contamination is due to oxygen; the figures for nitrogen and hydrogen are accurate to about ±10 p.p.m. and ±5 p.p.m., respectively, and the changes are probably not significant. Since the slice was 0·030 in. thick, a proportion (about 0·027 in.) of the parent sheet was included in the analysis, and therefore the true percentage of oxygen within the surface layers is a great deal higher than represented in Table I.

Oxidation of Titanium

The reaction between oxygen and titanium at elevated temperatures is presented here so as to clarify some observations made in the main body of this paper.

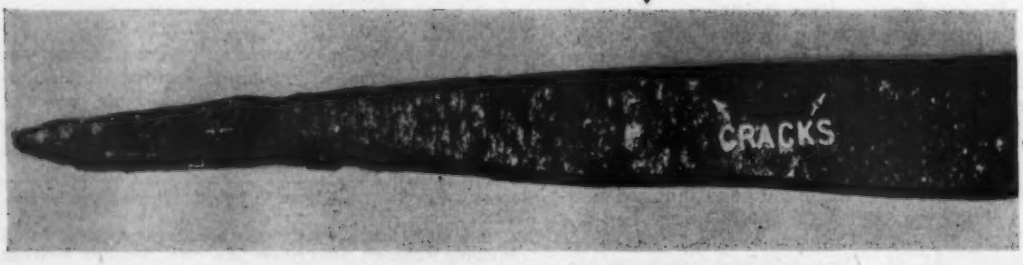
The reactions between titanium and atmospheric gases have been studied by various workers, and it has been established that oxygen and nitrogen will penetrate the metal lattice, causing severe embrittlement even when present in very small amounts.[2] With the exception of hydrogen, the gases react irreversibly and cannot be removed by vacuum heat-treatment.

As a result of his work on the oxidation of titanium, Jenkins[3] concluded that, for temperatures up to 1000°C., this process depends on the reaction between oxygen and titanium occurring at the metal/oxide interface. Above 1000°C. it is possible that diffusion of titanium ions outwards through the oxide scale occurs, altering the oxidation mechanism.

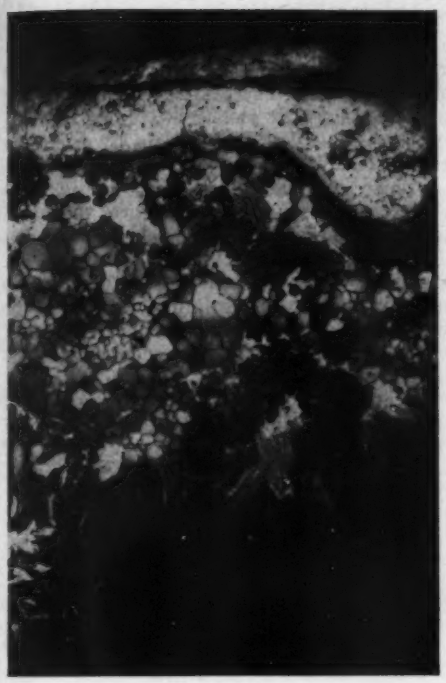


1—*Cut in 9 in. dia. billet of Ti–2·7% Cr–1·3% Fe alloy (Ti–150A), showing nature of cut face and location of specimen shown in Fig. 2*

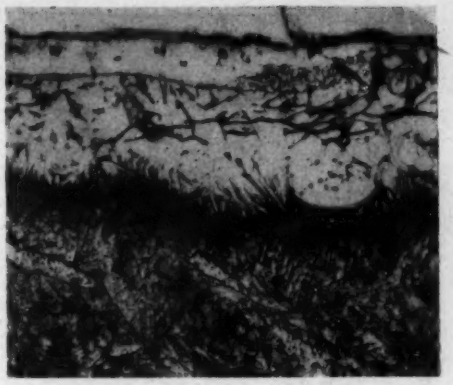
3—*Ti–150A: section showing detachable outer scale and porosity* ×10



2—*(Ti–150A): transverse section, showing crack and general effect of flame-cutting. Etched HF–HNO₃ solution* ×1



4—*Ti–150A: flame-cut section* ×500



5—*Commercially pure titanium: flame-cut section* ×500

The overall picture of oxidation processes taking place below about 1000°C. may be viewed as follows. The initial contact of oxygen with the metal surface produces a thin oxide film, through which further oxygen must pass so as to advance the state of oxidation. From theoretical considerations it can be deduced that the diffusion of oxygen through the oxygen-rich film is slower than that within the metal, so that it seems that the initial stages of surface oxidation are controlled by this process. In consequence of the difference in diffusion rates, an oxygen gradient is quickly established within the surface layers. Oxidation at this stage progresses by a joint mechanism involving further scale growth and oxygen solution in the metal, and when the scale reaches a critical thickness the rate of oxidation becomes linear. Jenkins also observed that if the supply of oxygen at the surface diminishes, then, as compensation, the metal absorbs part of the scale; it is thus concluded that the scale thickness at any instant is related to the oxygen gradient within the metal.

Metallographic Examination of Cuts

Specimen preparation

The sections selected for micro-examination were first ground down successively on silicon carbide papers of decreasing grit size, and finally polished on a 'Selvyt' cloth impregnated with gamma alumina moistened with an aqueous solution containing 0·5% HF. For general microscopical examination, the specimens were etched in 2% HF or in a solution comprising 4% KOH and 25 ml. H_2O_2 per litre. Before examining for cracks, all specimens were electro-polished, using an alcoholic solution of $AlCl_3$ and $ZnCl_2$.[4]

Oxidation products

Macro- and micro-examination of the samples revealed three distinct regions of contamination.

Detachable outer scale—The upper surface of the outer scale, which occurred only on the 9 in. dia. billet, was light brown or yellow in colour, and could be removed by hammering to expose a black adherent sub-scale. The general formation and brittle nature of the outer scale is shown in Figs. 1, 2 and 3.

Adherent scale—The black scale was present on all samples examined, and was strongly bound to the parent metal. Examination at high magnification revealed that it was composed of at least four layers, and Figs. 4 and 5, which show sections taken from the Ti–150*A* material and the unalloyed metal, respectively, are typical examples of the structures observed.

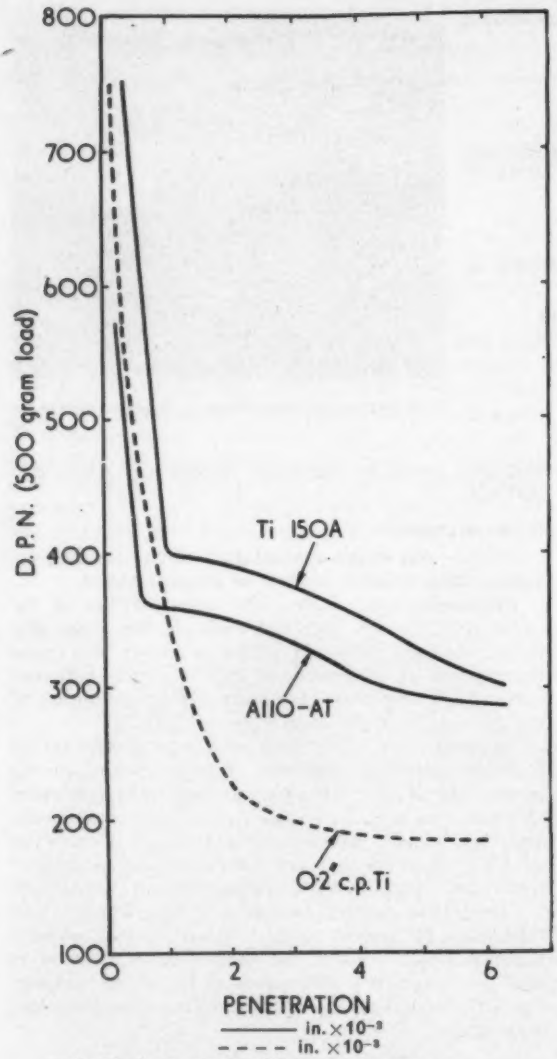
Core—This region is indicated in Figs. 4 and 5, and represents the parent metal core into which oxygen has penetrated. Where the oxygen concentration is high a blackening of the structure occurs on etching, but with decreasing concentration the colouring is less noticeable.

Defects

Major cracking in the underlying metal, as shown in Fig. 2, occurred only in the Ti–150*A* material, and is thought to be due to the hydrogen content of 150 p.p.m which at this level is likely to cause embrittlement and cracking in this alloy. Small fissures which occurred in the scale of all samples presumably resulted from growth stresses within the scale. The cause of the porosity shown in Fig. 3 is not known definitely. This region of the sample had been grossly overheated, and the material surrounding the pores reached the melting point, a condition which would not occur during normal cutting.

Hardness Surveys

Hardness surveys were conducted on the three materials, and the results are shown graphically in Fig. 6. The main hardening is restricted to less than 0·010 in. for the alloyed materials, and to about 0·002 in. for the commercially pure titanium. For the



6—Hardness surveys of flame-cut titanium

Ti–150*A* material the overall hardened zone extended to 0·060 in., and for the *A*110–*AT* to 0·045 in., but in the unalloyed material the hardened zone was much smaller, being about 0·003 in. The steep portion of the curves accounts for the main bulk of the oxygen absorption, whereas the remaining portion is accounted for almost entirely by transformation hardening. This was confirmed by conducting hardness surveys on welds made in the materials concerned.

Of interest also is the extent of hardening occurring at the outer surface of the 9 in. dia. billet. Hardness surveys in this region, which is shown in Fig. 2 as pronounced blackening of a triangular shape, revealed that hardnesses of up to 400 D.P.N. occurred at a distance of 0·12 in. from the surface.

A summary of the extent of contamination caused by the flame cutting is shown in Table II, together with the width of the hardened zone caused by rapid cooling.

Welding

Butt welding tests were made between cut edges of commercially pure titanium samples with the scale intact. Before welding, the sections were degreased in acetone, washed in alcohol, and thoroughly dried in a hot air stream. The welds were made in the B.W.R.A. chamber,[5] which contained high-purity argon, thus ensuring that no further contamination occurred during fusion. The welding speed was 3·5 in./min at 110 amp.

Examination of the welds showed that the oxides were completely taken into solution, and that the microstructure was no different from that of similar uncontaminated welds. This is not surprising, since the additional oxygen provided by the scale was certainly no more than 0·1%. A hardness examination of the weld metal showed that the increase in hardness for the $\frac{1}{8}$ in. sheet was 58 D.P.N., and that for the 0·2 in. sheet was only 38 D.P.N.

Discussion

The oxygen cutting of titanium involves preheating the titanium surface to the ignition temperature (about 1100°C.), so as to start combustion. Since the heat formed greatly exceeds the heat losses due to conduction, titanium cuts with great rapidity. For this reason the heating and cooling cycle can be made extremely short, thus minimizing the amount of contamination that occurs as the result of scale formation and oxygen solution in the metal. The nature of the cutting process ensures that the highly contaminated material formed is swept away. The amount of scale that remains on the edges after cutting depends largely on the thermal cycle that the material has undergone. A rapid heating and cooling cycle is beneficial, since a steep temperature gradient limits the overall penetration of oxygen to within the β region, and a high cooling rate minimizes the period in which the scale is allowed to build up.

The flame-cut samples showed that the depth of serious contamination was small. On the 9 in. dia. billet the total thickness of the scale was 0·01–0·03 in., and although this compares unfavourably with the smaller sections of the titanium–aluminium–tin alloy and the unalloyed materials, where the scale thickness is only 0·005 in. and about 0·002 in., respectively, the scale is porous and fragile and can be removed quite easily by hammering. The depth of hardening caused by the diffusion of oxygen into the metal core is much smaller, as shown in Table II.

Since the temperature of the cutting face is most probably of the order of the melting point of the material (≃1700°C.), it might be expected that a fair proportion of the oxide scale formed as a result of the cutting process is due to oxidation above 1000°C. However, in view of the steep temperature gradient existing at the cut surface, and the fact that the cooling cycle is extremely rapid, oxidation above 1000°C. probably contributes only a small part of the scale, and this may in fact be formed at temperatures below 1000°C. The results which are discussed seem to suggest that this is true for thin materials.

Table II

Extent of contamination in flame-cut titanium

Material	*Thickness, in.*	*Outer Scale Thickness, in.*	*Adherent Scale Thickness, in.*	*Core Contamination, in.*	*Hardness Zone Width, in.**
Ti–2·7% Cr–1·3% Fe alloy (Ti–150*A*)	9	0·01–0·03	0·005	0·003	0·050
Ti–5% Al–2½% Sn alloy (A110-AT)	3	Nil	0·003	0·002	0·040
Unalloyed titanium	0·20	Nil	0·002	0·001	<0·0005

* Due to transformation.

In summarizing, it can be said that the contamination arising from the flame-cutting process is restricted to the surface layers, and that its removal should not present any major difficulty. Removal of the outer scale is facilitated by reason of its brittle nature, as shown in Fig. 3; light surface grinding is sufficient to remove the major portion of the adherent scale.[1]

The major cracking in the $\alpha\beta$ material (Ti–150*A*) is probably due to high residual stresses in combination with hydrogen sufficient to induce embrittlement. It is possible, also, that cracking will be experienced in other $\alpha\beta$ alloys even when hydrogen is not present. In these cases cracking may be due to 'omega'-formation resulting from the partial breakdown of unstable β-phase. Caution is, therefore, advised in the application or subsequent treatment of $\alpha\beta$ materials which have been previously flame-cut.

The contamination of a butt-welded sheet 0·2 in. thick, which had been previously flame-cut, was small, and an increase in hardness of only 38 D.P.N. was recorded. For thicker sections the increase in hardness would be expected to be less, because of the increased weld-metal/oxide-scale ratio.

It is, of course, not recommended that welds between undressed edges be made, since in practice any increase in the oxygen level should be avoided. However, the low hardness increase in the weld metal shown by these tests has confirmed that the depth of contamination present on the cut edges is very small indeed.

Acknowledgments

The authors are grateful to Dr. A. A. Wells and Mr. G. Coates for discussion, and also to Mr. E. Booth of the U.K.A.E.A. (Woolwich Outstation), who carried out the gas analyses.

REFERENCES

1. G. COATES: *Engineer*, 1957, vol. 203, Jan. 1957, pp. 132–134.
2. A. D. MCQUILLAN and M. K. MCQUILLAN: "Titanium", 1st ed., p. 402. London, 1956, Butterworth Scientific Publications.
3. A. E. JENKINS: *J. Inst. Metals*, 1953–54, vol. 82, pp. 213–221.
4. A. D. MCQUILLAN and M. K. MCQUILLAN: *Loc. cit.*, p. 450.
5. A. MOMBRUN and W. G. HULL: *Welding Metal Fab.*, 1958, vol. 26, pp. 9–11.

Phase Analysis of Resistance Welded Joints in Unalloyed and Low-Alloy Carbon Steels

By D. Hrivňák and I. Hrivnák

By the use of electron microscope techniques a study has been made of the structure of welded joints made by the resistance welding process. Materials used ranged from low-carbon steel to duplex joints in high-carbon and austenitic stainless steels. The different structures obtained are illustrated in a series of micro-photographs.

Introduction

RESISTANCE welding is characterized by its high power requirements. Welding of the metal takes place within a period of a few cycles of normal alternating current, with a great, and sometimes even an enormous current discharge. By reason of its rapidity this method of welding is extremely economical and efficient. The short period of time and the high thermal gradient values create special conditions governing crystallization and phase transformations in the solid state. The heating rate may be partially characterized by the welding times, which on modern hydraulic resistance welding machines vary between 5 and 50 cycles (=1/50 sec). During the actual

Translated by M. de O. Tollemache for the British Welding Research Association from the Czechoslovakian periodical *Hutnické Listy*, 1958, vol. XIII, No. 3, pp. 233–241.

welding operation there is an increase in the compression force, and so the structure of the welded materials is influenced not only by thermal shocks, but also by strong dynamic factors. Under such conditions it is possible to identify in the structure of the welded joints various transitional and incomplete transformations, the internal structure of which has hitherto been unknown.

The present work covers a study of the nature of phase analysis of resistance welded joints in unalloyed and low-alloy carbon steels, and partly a study of carbon diffusion in 16% Cr–13% Ni austenitic steel in particular. For most of the material on which work has been done, there are no characteristic parameters, in accordance with which individual joints should be welded. Small changes in the parameters, moreover, do not have a fundamental influence on the structural composition.

This work forms only an introductory study of the phenomena and leads to a profound study of eutectoid reactions which will be presented by the authors in the near future.

The basic structure of unalloyed carbon steels may be ferrite and pearlite, pearlite, or pearlite and cementite. On welding these steels their structure undergoes transformation changes as a result of the influence of heat. Whilst the thermodynamics of the transformation on cooling have at least received partial consideration, the thermodynamics of the intensive and rapid heating have received little attention, and even less has been given to both parts of the thermal cycle. Since the heating is extremely intense, austenitization takes place with a very abrupt change in the concentration, and it frequently happens that the $\alpha \rightarrow \gamma$ transformation involves only a part of a crystal. Since in the α-phase the carbon concentration varies considerably (0·04% in ferrite and 6·67% in cementite) and carbon up to 1·7% is additionally soluble in austenite, austenitization must to a very great extent be accompanied by considerable carbon diffusion. And even though during heating the increasing temperature promotes this diffusion, the short period of the thermal effect is insufficient for complete equalization of the carbon concentration to occur. In the fusion zone, its proportions are more favourable to diffusion, and here greater equalization of the concentration likewise occurs.

Transformation in the solid phase may, according to Bowles and Barrett,[1] be divided into two types. One, the martensitic type, does not give rise to internal changes in the orientation of two neighbouring atoms, since the total movement of each atom relative to its neighbour is less than the atomic radius. Many observations show that the activation energy for the formation of the lattice during this transformation is only insignificantly compensated by the activation energy for diffusion at the same temperature. The martensitic transformation does not occur at a constant temperature, and when cooling ceases it comes to a standstill. That the absence of diffusion is characteristic of this transformation is confirmed by the fact that, while this particular phase has a superlattice, the martensitic phase also has a superlattice. In the other type, nucleation and the creation of a new phase take place as a result of the thermally activated diffusion of atoms from the existing phase to the new phase. This type of transformation is different from the former, in that nucleation and the formation of the new phase occur isothermally. As far as the structural state after resistance welding is concerned, both the martensitic and the pearlitic are characteristic.

The pearlitic transformation gives rise to very great carbon diffusion, since the concentration of carbon in the autectoid solution varies considerably. If a steel contains certain alloying elements (Cr, Mo, Ni), a longer time is required for diffusion, since their diffusion coefficients are of the order of 10^3 to 10^4 times less than the diffusion coefficient of carbon.

Experimental Section

Preparation of the specimens

In all instances a rapid metallographic method was used. After grinding to grain size 6/0 (Schröder) the specimens were electrolytically polished in a solution of the type $HClO_4+CH_3COOH+H_2O$, or $CrO_3+H_2O+CH_3COOH$, at normal current densities. To reveal the structure, 2% HNO_3 in ethyl alcohol was used, and only for specimens of duplex steel was use made of electrolytic etching in 10% H_2CrO_4 at 6 V with positive polarity.

For electron microscope examination various methods of preparing the specimens were used. Most frequently negative collodion replicas were used, which were shadowed *in vacuo* with up to 1 mg of Cr at an angle of 12–20°. In some instances polystyrene single stage prints were used. A print was obtained at 80°–150°C., by moderate pressure of the surface under examination against the polystyrene layer, placed on a glass underlay. The specimen was separated from the polystyrene at reduced temperatures (5°–10°C.). *In vacuo* a thin layer of chromium was vaporized onto the print at an angle of 12–20°, and the shadowed carbon surface was vaporized at right angles. Then the polystyrene was dissolved in benzene, and the replicas obtained were washed repeatedly. The carbon extraction replicas were obtained by Smith and Nutting's method. Onto the polished, etched, and cleaned surface a thin layer of carbon was vaporized *in vacuo* at a moderate angle, and this was done on a vertical stand with a distance of 15–20 cm between the surface of the specimen and spectrographically pure carbon electrodes. Good results were obtained at 24 V and a current of 40–50 amp. The time of vaporization varied between 4 and 8 sec. The carbon film was separated from the surface in 10% HNO_3, then washed in 10% HCl, and in a number of water baths.

For the X-ray investigations cobalt radiation was used with a wave length of $\lambda_{CoK\alpha\beta}=1{\cdot}785287$ Å. Silver was used as a reference line. For high measurements the reverse reflection method was used. Microhardness measurements were effected by Hanneman's method at a 50 g load on a Zeiss-Neophot microscope.

Phase analysis of resistance welded joints in low-carbon steels

A number of welds in soft, unalloyed steels with a carbon content of 0·12–0·2%, obtained on a resistance welding machine, were subjected to electron microscope examination. It is obvious that the strength of a joint is considerably influenced by its homogeneity, as well as by its structural state. This work does not deal

with the strength of the weld, but only analyses the phase changes that occur during the welding process.

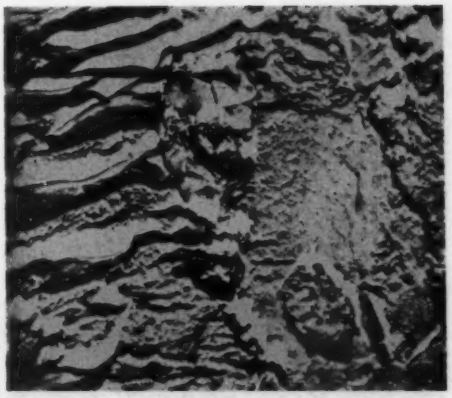
The basic material for all the specimens was ferrito-pearlitic, with a structure oriented in the direction of rolling. The heat-affected zone was first of all marked by the pearlitic retransformation from the A_1–A_3 range of temperatures. During the rapid heating, austenitization of the pearlitic grains occurred, and carbon from these areas was diffused into the surrounding ferrite. Such carbon diffusion occurs primarily on the grain boundaries and along other clearly defined lines where the maximum number of vacant sites exist. From these regions the carbon diffusion also spreads to the ferritic grains which are progressively transformed into austenite. During the rapid cooling which follows the short heating period, the diffusion process is variable, and the $\gamma \rightarrow \alpha$ retransformation is not completed, so that the structure does not revert to its state before welding. For subsequent comparison, in Fig. 1 is shown the different orientation of the pearlitic grains of the base metal area, which are not affected by the heat. Here it is a question of lamellar pearlite, the lamellae of which are distinguishable even at magnifications of ×500–750. On the other hand Fig. 2 characterizes the zone of pearlite retransformation where, as a result of the short thermal shock, selective infiltration occurs on the boundaries of the pearlite and ferrite. The fundamental pearlitic nucleus preserves its coarse lamellar structure, whilst the areas which have undergone the $\alpha \rightarrow \gamma \rightarrow \alpha$ transformation have a different orientation, which is characterized under an optical microscope as 'soft, disintegration phases'. This is a carbon-saturated α-phase, in which the carbide is distributed in very fine lamellar form, very much smaller than that of the original carbide texture.

The micro-hardness of this phase, which qualitatively does not differ from the eutectoid pearlitic transformation, is somewhat less (200–300 D.P.N.) than the optical micro-hardness of the lamellar pearlite (300–400 D.P.N.).

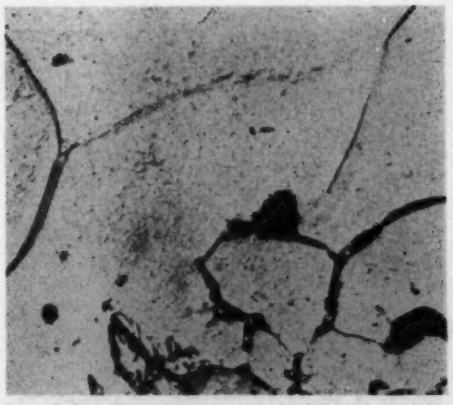
The zone of the retransformed pearlite is shown in Fig. 3; there is not only carbon diffusion on the boundaries of the ferritic grains, but also an extremely fine lamellar form of the pearlite. Since this is a reproduction of an extraction print, it shows the form of the Fe_3C carbides as originally distributed.

In the further zone affected by temperatures above the A_3 point full austenitization of the material takes place. And even though conditions were favourable to carbon diffusion, complete equalization of the carbon concentration occurred only in the region of high austenitization temperatures. During cooling an eutectoid pearlitic transformation only rarely occurs, and for the most part in such areas soft and hard disintegration phases are formed with Widmanstätten orientations. In these disintegration phases it is a question of saturated ferrite, in which finely dispersed lamellar carbide particles are distributed in definite preferential directions. Figure 4 gives a characteristic and striking example of the Widmanstätten orientation of a disintegration phase with carbides precipitated in step-like layers. It is interesting that a superstructure is formed in the shape of a network of carbides on the freely precipitated ferritic grains (Fig. 5). This probably results from the diffusion of carbon into the step-like layers, where the carbide phase is formed on cooling.

1—*Detail of pearlitic crystal. Polystyrene—C–Cr replica. (2% HNO_3)* ×7500



2—*Zone of retransformed pearlite. Collodion—Cr. (2% HNO_3)* ×7500



3—*Zone of retransformed pearlite. Carbon extraction replica. (2% HNO_3)* ×7500

The possibility cannot, however, be excluded that it is a question of tertiary cementite, precipitated at temperatures below the A_1 point.

By optical microscope examination it is possible to discern in the zone affected by temperatures above the A_3 point, firstly refinement of the grain, and in addition some partial coarsening, as transition of the structure occurs towards that of the metal fusion zone. This seems to be a question of partial arcing in the weld as a result of oxidation of the abutting surfaces. It is indeed true that during resistance welding oxidation of the abutting surfaces does occur, but this could not be the cause of such extensive arcing. There are two reasons against this: (*a*) It is an extremely short process, during which oxidation could be the cause of arcing in only a small area of the fused metal; and (*b*) oxidation is removed from the weld by, for instance, planing. From electron microscope examination it is possible to explain this apparent arcing as a qualitative transformation in the structure of the precipitated carbide, which at these points is precipitated in extremely fine lamellar form, and thus causes less diffusion of optical light. Figure 6 indicates the fusion zone with such very finely laminated textures of the disintegration phase.

Measurement of the micro-hardness did not show any marked variations, and apart from a gradual increase in the transition zone, which had no material influence on the mechanical and physical properties of the weld, the hardness pattern was not marked by any anomalies.

It is often difficult to determine whether a resistance welded joint is the result of fusion or diffusion. From this viewpoint characteristic signs of fusion joints were observed. On the basis of several experiments it became clear that a good criterion for distinguishing the existence of fusion in a joint in plastically deformed materials is an Oberhoffer macro-etching. This reagent discloses segregation and reveals the ridge-like structure of the material. The ridge-like structure of the material can only be removed by fusion. Thus, for instance, a good indication of a fusion joint is the disappearance of the ridge-like structure in the weld-metal area. In diffusion joints it is possible to observe deformation of the ridge-like structure only in the direction of flow of the metal. Another characteristic sign of a fusion joint is the manner in which the crystallization occurs around the oxide compounds in the area of the joint. It is known that inclusions and oxides form crystalline nuclei, around which ferrite is precipitated. Where the grains of the material (ferrite) are distributed regardless of the line of the oxides in the area of the weld, a fusion weld is formed. But where this disposition strictly limits the grains on one side or the other, then it may also be a diffusion joint. The pattern of the micro-hardness is not a good criterion for the determination of the existence of fusion in a joint, since it only gives a picture of the structural transformations, regardless of their primary crystallization. Although there is some sort of difference between fusion and diffusion joints, and it is also small, it is still within the ranges of the distinguishable values, obtained by measurement.

From X-ray micro-analyses it was clear that the lattice parameter over the whole area of the joint and the heat-affected zone is essentially unchanged, and only a gradual drop in the intensity of the blackening of the Debye lines in the direction of the weld indicated this zone. This drop may be caused by tertiary formations.

5—Ferrite with a superstructure. Collodion. (2% HNO_3) ×7500

4—Widmanstätten structure. Collodion. (2% HNO_3) ×7500

6—Metal fusion zone. Collodion. (2% HNO_3) ×7500

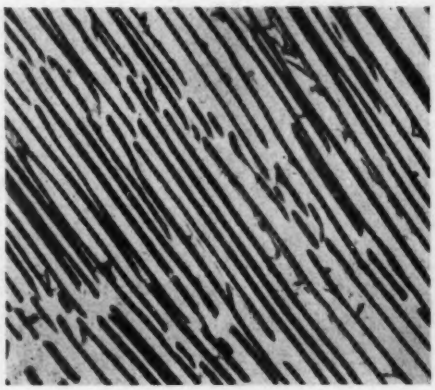
Phase analysis of resistance welded joints in unalloyed medium carbon steels

It is clear that during very rapid welding (more rapid than the cooling) austenite cannot be transformed into the disintegration phases of a soft texture with carbide inclusions, but in the grains martensitic transformations can occur without diffusion. A localized zone of a material, which happens to be in the austenitic state (*e.g.*, a nucleus of a pearlitic crystal), can during rapid cooling lead to the thermal hardening of this area in the surrounding ferrito-pearlitic region. This feature occurs relatively frequently in resistance welded joints. For materials with a high content of carbon, the pearlite content will also be higher, and the disintegration phase will be harder. But there will also be a greater risk of local thermal hardening, as well as of thermal hardening of a zone heated to the austenitic state. A good example of such a joint is a resistance weld in unalloyed 0·6% C steel. The base metal is pearlitic with a narrow ferritic fringe on the boundaries of the grains. The configuration of the pearlitic nucleus, unaffected by heat, is shown in Fig. 7.

The material examined was relatively coarse grained, and the eutectoid temperature which separates the austenitic and ferritic zones during welding had affected half the pearlitic grain. As a result of the rapid cooling this zone was transformed without diffusion with a high degree of hardness (up to 1100 D.P.N.). Figure 8 is an electronographic print of the boundary between the untransformed pearlite and ferrite and the product of transformation without diffusion, which has preserved its original orientation. But nearer to the weld in this structure the carbides start to precipitate into lamellar formations.

Up to the present time the mechanism of martensitic transformation has not been unequivocally described neither under conditions of rapid cooling, nor under those of rapid heating, which combined with high pressures during the welding process may lead to divergence from the process as observed in the laboratory. It is also illuminating that three phases can exist side by side with different properties—soft ferrite, which even in martensitic surroundings forms the boundary of the former α-grains (Fig. 9);* pearlite, which as a result of the abruptness of the thermal shock and the low, sub-eutectoid temperature, did not undergo transformation; and the product of retransformation without diffusion, which has no martensitic contours. This phenomenon can be attributed only to the fact that the thermal shock caused an isothermal temperature effect, since the eutectoid temperature (721°C.) did not affect even half of the pearlitic grain. It must be pointed out that this was not a static or quasi-static effect, but a dynamic thermal shock, which to such an extent localized the transformation zone that on the boundary of the unaffected pearlite there was no gradual transition in the concentration. The existence of a localized area of the original ferrite in a zone of transformation without diffusion may be explained by the fact that, at the low temperature and during the short time for carbon diffusion, austenitization could not take place at these points. This joint, of which the mechanical properties are completely unsatisfactory, is at the other extreme of the whole range of structures which occur in resistance welded joints in unalloyed carbon steels. Joints which have satisfactory mechanical properties have structural orientations similar to those previously described.

7—*Pearlite. Carbon extraction replica. (2% HNO_3)* ×*15000*



8—*Ferrite–pearlite–martensite boundary. Collodion. (2% HNO_3)* ×*7500*

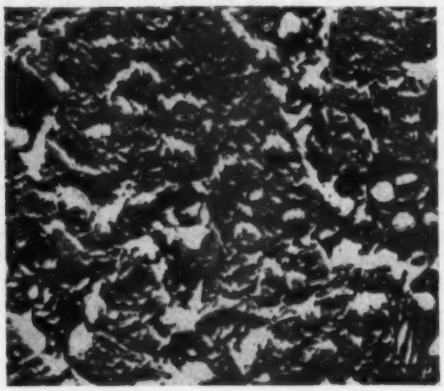
* This photograph is from the same material, and it was kindly provided by P. Slyško.

9—*Zone affected by transformation without diffusion. (2% HNO_3)* ×*350*

10—*Coagulated carbides with crystalline sub-structure. Collodion. (2% HNO_3)* ×15000

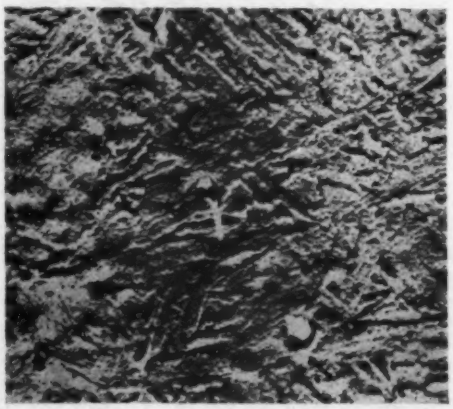
Phase analysis of resistance welded joints in high carbon steels

All the ranges of structures found in resistance welded joints in unalloyed and low-alloy carbon steels are difficult to cover exhaustively in this short treatise; in addition as an interesting example of phase analysis a resistance welded joint in high carbon steel (1·19% C, 1·24% Cr) with austenitic material of the type 16% Cr–13% Ni will be discussed. The basic material was ferritic with spheroidal carbide formations, whose structural nature was impossible to distinguish with the optical microscope even at the highest magnifications. From electronographic analysis it became clear that it was a matter of peltate* formations, which at sub-eutectoid temperatures are transformed into spheroidal formations (Fig. 10). The investigation of the sub-structure of these carbides is extremely interesting. From the structure of the sub-lattice, as from many other experiments also, it becomes apparent that in no instance is it a matter of an artefact which could be caused by the technique of producing the prints. One author has recently observed[23, 24] the crystalline sub-structure of an intermediary σ-phase in austenitic 16% Cr–13% Ni steel, which was probably developed out of carbides of the type $(Fe, Cr)_{23}C_6$. This other example of the occurrence of a crystalline sub-structure may be evidence of similar phenomena, since the spheroidal particles in Fig. 10 are likewise carbides of Cr. The low micro-hardness of the basic structure (~200 D.P.N.) refutes any idea that this ferritic structure is saturated by carbon. In the outer, heat-affected zone fundamental changes occur in the configuration and the actual structure of the carbide phase. This is dissolved in the austenitic matrix, and on cooling has a tendency to form lamellae. Figure 11 shows the tendency towards a lamellar structure of the carbides, which, however, are distributed as extremely finely dispersed particles in the saturated ferritic matrix. The spheroidal formations of the chromium carbides are difficult to dissolve and retain their crystalline structure. Nearer to the weld greater homogenization of the austenite occurred, and on cooling the carbides are precipitated from the ferritic phase in a fine form. Here, however, it is also possible to distinguish the original areas of carbides, which still do not have a substructure. The whole orientation of the carbides reminds one of the Widmanstätten structure, which is extremely close to the martensitic structure (Fig. 12). In the first series of experimental welded joints the welding technique was not sufficiently developed to obtain satisfactory mechanical and



11—*Heat-affected zone. Collodion. (2% HNO_3)* ×7500

12—*Hard disintegration phase. Collodion. (2% HNO_3)* ×7500



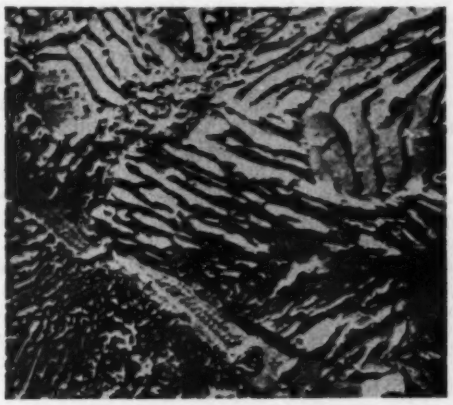
13—*Martensite with fine striation. Collodion. (2% HNO_3)* ×7500

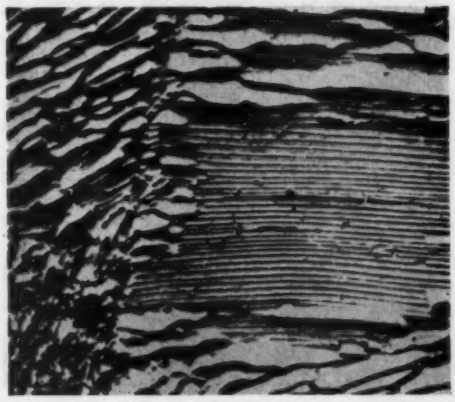
* *Translator's Note.*—These have been referred to in English texts as 'plate-like', but the Latin or Greek shield (pelta) is a more accurate description and translation of the original.

structural values for the joints. In all instances there occurred in the transition zone a hard martensitic structure without striation, or with fine striation. Figure 13 illustrates this structure with fine martensitic striation, in which is visible the beginning of the precipitation of the carbide structures. The micro-hardness of this structure attains 1100 D.P.N. in the case of diffusion joints and 600 D.P.N. in the case of joints with a fusion character. The lower hardness of the martensitic phase, where other conditions remained equal, may be explained by carbon diffusion into the austenite (such diffusion was revealed), which was considerably greater in the fusion joints than in the diffusion joints. A decided part may also be played in this by the accumulated energy of the latent heat of solidification, which could moderate the intensive cooling.

To determine the character and extent of the carbon diffusion into the austenite, for each of the diffusion joints, heating was carried out for 150 hr at a temperature of 650°C., with subsequent cooling in the furnace to 300°C. and thereafter in air. In the product resulting from this heat-treatment there was carbon diffusion from the carbides to the ferritic matrix, where there was formation of coarse-grained areas with an optically visible sorbitic character. The grain boundaries were fringed with a carbide substance, the hardness of which could not be determined owing to its small dimensions. It was, however, greater than the hardness of the sorbitic disintegration phase. Electronographic examination, however, revealed lamellar structures in the disintegration phase with very fine pearlitic structures. The carbide substance surrounding the sorbitic grains again had a crystalline sub-structure, from which it is possible to judge its identity with the carbide structure of the preceding analysis. It is interesting to study the morphology and structure of this crystalline sub-structure, which has, so it appears, a fairly regular hexagonal structure (Fig. 14). In all instances, indeed, this was observed at a given distance from the boundary with the neighbouring phase. But not only is the structure of the pearlitic (sorbitic) grains heterogeneous, but in the structure it is also possible to discern various anomalies to which it is difficult to give interpretations. Thus, for instance, Fig. 15 shows dual crystallization of cementite lamellae with 'coarse' and very fine structures. Most interesting of all is the revelation that between the coarse lamellae it is possible to discern the crystallization of finely laminated foliates. After prolonged heating, an extremely fine pearlitic (sorbitic) substance was formed out of the original martensitic zone, and this substance had lamellae which were much less oriented in the basic direction (60°). In the transition zone this resulted in intensive carbon diffusion into the austenite. This decarbonization resulted in the formation of free ferrite alongside the pearlite. It is possible that this may be chrome ferrite, for it is unlikely that chromium diffuses into the austenite (see Fig. 16). Interesting also is the electronographic study of the boundary between the carbon and austenitic steels. As a result of the carbon diffusion it is impossible to establish without ambiguity that the boundary in Fig. 17 is austenitic; rather it may be a localized ferritic area with extremely finely precipitated carbides. In this 'neutral area' the carbon diffusion is brought about only on the boundaries of the γ-grains and in other directions where there is intense precipitation of carbide particles without distinct regular orientation. The morphology of this probably already austenitic

14—*Orientation of sub-structure of chromium carbides. Collodion. (2% HNO_3)* ×7500





15—*Anomaly in structure of pearlitic grain. Collodion. (2% HNO_3)* ×7500

16—*Ferrite in transition zone. Collodion. (2% HNO_3)* ×7500

area is indicated by Fig. 18. In the further diffusion area, precipitation of chromium carbides occurred on the boundaries of the γ-grains with its characteristic crystalline substructure (Fig. 19).

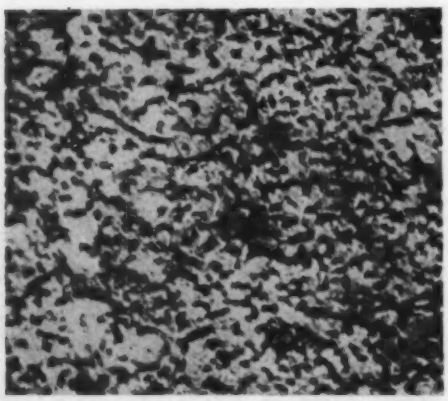
Discussion

In optical metallography it is customary to make qualitative discrimination between the products of eutectoid transformations according to their appearance at definite optical magnifications. Thus, for instance, the lamellar structure of pearlite is initiated by the growth of carbides with lamellar structures, between which are lamellae of ferrite. Troostite is formed during rapid cooling in the eutectoid temperature range, and optically it has no lamellar structure. In addition to this difference, there also occur differences in their physical and mechanical properties, such as strength, density, macro- and micro-hardness. These differences are so great that by them it is possible to define the various structural components. At the large magnifications obtained electronographically, however, it is possible to determine that it is a matter of lamellae, the only difference between them being the coarseness of the lamellae and the distance between them. Thus, for instance, the structure shown in Figs. 1 and 2 is pearlite, but the structure shown in Figs. 14 or 15 is also pearlite. The only difference is the ten-fold refinement of the lamellae, which is impossible to distinguish optically. It is an established fact that the mechanical properties of both formations are different. It must, of course, be remembered that in the study of mechanical properties and their comparison, the problem is not approached from the aspect of a comparable structure, but from that of a comparable composition. Thus, for the coarse lamellar pearlite, properties are obtained for a pearlite-ferrite complex, whereas in the disintegration phases the composition is more nearly homogeneous. Qualitatively the disintegration phase shown in Fig. 7, for instance, differs from pearlite. And where there is a very much finer lamellar structure, the cementite lamellae are extremely thin and do not have a linear orientation. A striking indication is given by the discovery that in the disintegration phases with a Widmanstätten character, elementary carbide lamellae are also precipitated at an angle of 60° or even 120°, and these linear lamellae lie within the limits of 0·1–0·15 micron. The whole structure has the appearance of saturated ferrite, because even though its carbide lamellae are extremely fine (less than 0·01 micron), the interlamellar spacings are many times greater (up to 10 times).

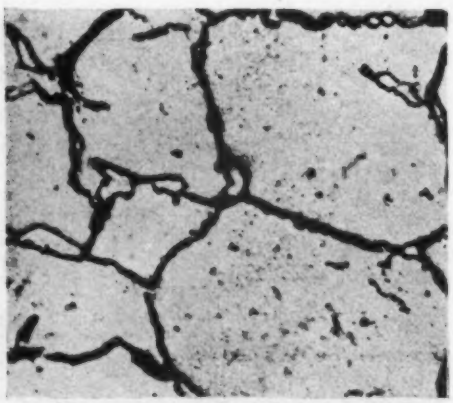
The structure of the pearlitic grains is also extremely complicated, and the anomalous phenomenon of the dual transformation, evidenced by Fig. 15, may also be explained by the fact that by precipitation of the chromium carbides the area surrounding them is enriched in this component. During cooling there took place a pearlitic transformation, initiated by the iron carbides Fe_3C, and the ferrite,enriched with chromium, filled the interlamellar spacings. With the further reduction in temperature there is formation of complex carbides $(Fe, Cr)_2C$, which are precipitated in a sub-lamellar form with a comparable orientation.

Likewise the crystalline sub-structure of the chromium carbides, as shown in Figs. 10 or 14, is singular, and likewise its regular lattice structure. As has been mentioned, this sub-structure was observed during the formation of the σ-phase in austenitic steel

17—*Transition zone. Collodion. (10% H_2CrO_4 at 6 V)* ×7500



18—*Another area of carbon diffusion. Collodion. (10% H_2CrO_4 at 6 V)* ×7500



19—*Carbon diffusion on grain boundaries. Collodion. (10% H_2CrO_4 at 6 V)* ×7500

(16% Cr–13% Ni).[23, 24] To reveal this, originally electrolytic etching in 10% NaOH at 6 V was used, but in this instance the existence of the substructure was revealed by a 2% alcoholic solution of HNO_3, and in some cases by electrolytic etching in 10% H_2CrO_4 at 6 V. Since three different reagents were used, there can be no doubt that the crystalline substructure forms part of the internal structure. In some instances it is possible to observe a regular hexagonal structure of the crystalline elements. Therefore two explanations are possible: (*a*) Either the crystalline sub-structure is peculiar to a certain type or types of chromium carbide; or (*b*) the crystalline structure is characteristic of a certain internal orientation, .a certain crystallographic composition without regard to the elements which crystallize in it. But it is not excluded that both explanations are possible. The revelation of a crystalline sub-structure in the carbides portrayed in Fig. 19 strongly supports the first alternative. Among the physical properties of this phase only its hardness, which was appreciably higher than 400 D.P.N., was discovered.

Conclusions

In the present work, phase analysis was carried out on joints obtained by resistance welding in various groups of steels. Joints in unalloyed, low-carbon material with 0·12–0·15% C, a resistance welded joint in medium-carbon steel (0·6% C) with a thermally hardened zone, and duplex resistance welded joints in high-carbon, austenitic material, were chosen as typical.

It was discovered that various transition and disintegration structures are distinguishable with an electron microscope by the manner of precipitation, and the form, of the precipitated carbides. Even for a product of a transformation without diffusion, all the carbide formations are lamellar and qualitatively (in essence) indistinguishable. The softest disintegration phase is ferrito-cementitic by reason of its orientation and structure, and by reason of the lamellar carbides. In these latter phases, however, there is a variation in the proportions of the coarse lamellar formations and the interlaminar ferritic filler. Whereas in pearlite this proportion is 1 : 6, in the latter instance the relationship is 1:10 in favour of ferrite. The discovery is also important that, whereas the complete structure has a Widmanstätten character, the basic carbide lamellae form a serrated line with an angle between the teeth of 60° or 120°. This applies likewise to isolated precipitation from the saturated ferrite.

In the case of resistance welds in steel with 1·19% C and 1·24% Cr an interpretation was given of electronographs of the phase composition of high-carbon, low-alloy steels.

The authors draw attention to the crystalline substructure of the chromium carbides of the type $(Cr, Fe)_2C$. The transitional carbides which arise during the diffusion of carbon into the austenite are morphologically interesting.

BIBLIOGRAPHY AND REFERENCES

1. J. S. Bowles and C. S. Barrett: *Progress in Metal Physics*, vol. III, pp. 1–41.
2. Umanskij, Finkelstein and Blanter: Budapest, 1952.
3. R. F. Mehl: *J. Iron Steel Inst.*, 1948, vol. 159, pp. 113–129.
4. B. J. Lyubov: *Zh. Tekhn. Fiz.*, 1950, vol. 20, No. 7, pp. 872–879.
5. B. J. Lyubov: *Dokl. Akad. Nauk. SSSR.*, 1948, vol. LX, p. 795.
6. B. J. Pines: *Zh. Tekhn, Fiz.*, 1948, vol. 18, p. 831.
7. A. Schrader and F. Vever: *Arch. Eisenhüttenwesen*, 1952, vol. 23, pp. 489–495.
8. A. J. Gardin: *Zh. Tekhn. Fiz.*, 1954, vol. 24, pp. 686–693.
9. G. V. Kudryumov and M. D. Perkas: 2nd Symposium of Monographs of SNIIChERMET, Metallurgizdat, 1952.
10. A. Hultgren *et al*: *Rev. Mét.*, 1953, vol. 50, No. 11, p. 737.
11. J. Čadek: *Hutnické Listy*, 1957, vol. 12, No. 7, p. 585; No. 8, p. 687.
12. Le Claire: *Progress in Metal Physics*, vol. 1.
13. *Idem.* vol. IV.
14. B. D. Pines: *Zh. Exp.i Teor. Fiz.*, 1943, vol. XIII, p. 411.
15. B. H. Alexander and F. N. Rhines: *J. Met.*, 1950, Oct., p. 1267.
16. G. Wells, W. Batz and R. Mehl: *J. Met.*, 1950, Nov., p. 1375.
17. J. D. Bakalar: *J. Met.*, 1950, Oct., p. 2317.
18. F. S. Buffington: *J. Met.*, 1950, Oct., p. 2317.
19. L. C. da Silva and R. Mehl: *J. Met.*, 1951, Feb., p. 155.
20. A. A. Smirnov: *Zh. Exp.i Teor. Fiz.*, 1947, vol. 17, p. 730.
21. D. Hrivňáková: Diploma treatise, Bratislava, 1957.
22. I. Hrivňák: *Zvaranie*, 1957, No. 3.
23. I. Hrivňák: *Hutnické Listy*, 1958.
24. I. Hrivňák: *Zváračsky Sbornik SAV*, 1958, No. 2.

STOP PRESS

Exhibition of New Welding Methods

The South London Branch of the Institute has arranged to hold a small exhibition of "Some Modern Methods of Welding" at 54 Princes Gate, London S.W.7, on Tuesday, 2nd June.

Exhibits will include a number of types of equipment for fine wire welding of aluminium, and machines for ultrasonic welding and for welding reactive metals by the electron-beam process. It is also hoped to show specimens of components fabricated by these different techniques. The exhibition will be open from 3.0 p.m. to 8.0 p.m. and tea will be available. All members are invited to attend and to bring their friends.

NEWS
and Announcements

News of the Institute and Branches
B.W.R.A. · Other Societies ·
and Industry

INSTITUTE ACTIVITIES

Council Nominations

At its meeting on 12th March, the Council unanimously nominated the following candidates for the Presidency, Vice-Presidency, and Honorary Treasurership:

President E. Seymour-Semper, M.I.Mech.E.

Vice-Presidents W. Barr, O.B.E.
E. Fuchs, M.A., A.M.I.Mech.E.

Hon. Treasurer W. E. Harriss

Benefactors

The South London Branch Committee has presented to the Institute an oak cupboard for the Council Room and Lecture Hall, matching the furnishings.

A.E.I. Limited have presented lighting fittings for the Council Room and Lecture Hall.

The Council offers its warm thanks on behalf of the Institute for these generous gifts.

Meetings

Autumn Meeting 1959

The Autumn Meeting for 1959 will be a joint one with members of the French Institute of Welding. The theme of the papers will be "Quality Control in Welding". The tentative programme is as follows:

Tuesday 3rd November:
Morning and Afternoon—Works Visits
Evening —Reception at Institute and Presidential Address.

Wednesday 4th November:
Morning—Works Visits
Evening —Annual Dinner at Park Lane Hotel.

Thursday 5th November:
Morning and Afternoon—Technical Sessions at the Institute.

Friday 6th November:
Morning—Technical Session at the Institute.

Slag Welding

A half-day Symposium on "Slag Welding" will be held at the Institute on the afternoon of **Thursday 1st October 1959**, when it is hoped to present and discuss a number of papers on recent developments in this Russian welding process.

PUBLICATIONS

Inert-Gas Arc Welding

The Institute's Inert-Gas Arc Welding Panel is at present compiling a Handbook on Inert-Gas Arc Welding. It will be in two parts, and the draft contents are:

Part I: Review of available processes; pictorial review of applications; safety precautions.

Part II: Aluminium and its alloys; nickel and its alloys; copper and its alloys; corrosion- and heat-resisting steels; magnesium alloys; carbon steels; titanium and its alloys, zirconium, tantalum, and other special materials.

The membership of the Panel is as follows:

Chairman: R. R. Sillifant (British Oxygen Engineering Ltd.)

M. Birkhead (W. P. Butterfield Ltd.); J. C. Borland (B.W.R.A.); N. T. Burgess (Aluminium Development Association); J. Hinde (Mond Nickel Co. Ltd.); J. E. Hooper (Bristol Aero-Engines Ltd.); P. T. Houldcroft (B.W.R.A.); J. A. McWilliam (Firth-Vickers Stainless Steels Ltd.); D. Slater (A.P.V. Co. Ltd.); D. B. Tait (Quasi-Arc Ltd.); C. A. Terry (I.C.I. Ltd.); H. F. Tremlett (B.W.R.A.); D. J. Whitehead (Magnesium Elektron Ltd.); W. A. Woollcott (British Oxygen Gases Ltd.).

Site Welding of Building Structures

The Site Welding Panel of the Structural Sub-Committee is now working on the final draft of a booklet on the Site Welding of Building Structures.

The purpose of the booklet is to encourage the wider application of welding on site through a better understanding of the correct approach to design, fabrication, and erection.

Index to Transactions and Welding Research, 1938–1953

The Institute has published a comprehensive Index, under subjects and authors, to the 16 volumes of its *Transactions* issued between 1938 and 1953, and to the 7 volumes of *Welding Research*, published for the British Welding Research Association between 1947 and 1953.

Most of the results of the researches undertaken by the Welding Research Council of the Institute and by the Research Association before 1953 were published in these two periodicals. The combined Index therefore affords ready reference to an important body of material, and should be valuable in any technical library, whether it possesses copies of the periodicals or not.

The price of the book is 25*s.* plus 1*s.* postage, but Members of the Institute and of the BWRA may obtain copies for 20*s.* plus 1*s.* postage. Orders should be sent to the Secretary, Institute of Welding, 54 Princes Gate, London, S.W.7.

NEWS OF MEMBERS

Mr. **R. C. Thomson**, chief welding instructor in the Training Department of G. A. Harvey & Co. (London) Ltd., has been appointed Company Welding Engineer.

Dr. **W. I. Pumphrey**, Manager of the Research Department of Murex Welding Processes Ltd., has been elected to the Court of Governors of the University of Birmingham, as a representative of the Guild of Graduates.

Mr. **G. Cubitt-Smith**, Senior Planning Engineer with Crompton Parkinson Ltd., has been appointed Production Engineer at the Chelmsford Works of the Company.

Mr. **W. P. Seymour**, founder-member of the Tees-side Branch of the Institute 25 years ago, has retired after more than 50 years with W. T. Seymour Ltd., established by his father in 1880.

At the meeting on 12th March, the Council was glad to learn of the recovery of three of its members:

Mr. **A. J. Francis** of the South Western Branch, who has recently undergone an operation for cataract.

Mr. **A. H. Scothern,** the representative of the East Midlands Branch, who is convalescing after a long attack of sciatica.

Mr. **E. S. Waddington,** who underwent an operation for the removal of the gall-bladder, and is also convalescing.

The Council sent its best wishes to all these members.

Obituary

The Council regrets to announce the deaths of:

L. C. Shervell (South London Branch, Member 1944).

T. R. Das Gupta (Indian Branch, Associate 1958).

Charles H. Bennett, attached to the Liverpool Branch, who was killed in a motoring accident on 3rd March (Associate, 1951).

BRANCH NEWS—Reports of Meetings

Liverpool — Non-Ferrous Welding

At a meeting in Liverpool on 10th February, Mr. E. A. Foreman gave an instructive lecture on the "Welding of Non-Ferrous Metals".

The largest group of these comprises the aluminium alloys, which have been increasingly used in almost all industries in recent years.

Aluminium is fairly easy to weld, and the joints are strong and have the same characteristics as the parent material. Except for the heat-treatable alloys welding has no detrimental effect and no elaborate treatment after welding is needed although with gas welding thorough removal of flux residues is essential.

Some information on the properties and uses of fluxes and on standard welding equipment was given by Mr. Foreman. He then went on to describe methods for welding castings. These lend themselves readily to fusion welding, the main precaution being the complete removal of grease, oil, or atmospheric corrosion. Aluminium castings, like those in other metals, generally need to be preheated. Some care is needed however, for aluminium can overheat very quickly, and with little warning, so that the casting can collapse under its own weight.

Aluminium is also joined by brazing methods. These have the advantage of needing a lower flame temperature so that light components and thin edges can be joined without fear of them melting away.

Mr. Foreman then gave similar information on copper. This is a metal that can be joined by almost any process; it can be hard or soft soldered, brazed, bronze welded, or fusion welded by gas, metal-arc or inert-gas metal arc processes. Since copper has a high afinity for oxygen at fusion temperatures de-oxidized material is the most suitable for welding purposes. Mr. Foreman described some of the commercial grades that are available and provided useful information on various types of welding rod and fluxes.

Since copper has a high thermal conductivity it is necessary to supply more heat during welding or brazing than for other metals such as steel or aluminium. In gas welding the flame adjustment is critical. The need for after treatment is not yet generally agreed, but sometimes a light hammering of the joint is given when the material is still above 600°C.

With most non-ferrous materials the need to use flux and the consequent difficulty in the removal of flux residues has led to an increasing use of the processes carried out under a shield of inert gas. To illustrate the applications of inert-gas metal-arc welding the lecturer showed the sound colour film "Argonaut Welding". He then undertook several experiments, in which a number of industrial applications of oxygen were demonstrated.

A vote of thanks to the speaker was given by Mr. Tachell, the Hon. Secretary of the Manchester Branch. H.R.C.

Successful Salesmanship

The theme of the lecture at the meeting on 10th March took a course different from those usually associated with welding practice, when Mr. R. H. Boughton spoke on "Welding Equipment Salesmanship".

At the beginning, Mr. Boughton explained that he thought salesmanship covered a far larger sphere than that generally associated with salesmen. The whole of the commercial organization manufacturing and supplying welding equipment should really be considered as a part of the sales organization, for selling was the creation of a demand for the products and the ability to deliver them to the customer at the right price and at the right time. To ensure that this could be done in future it was necessary for the whole organization to work as a team and to look ahead to see what improvements could be made to existing designs, what types of new equipment were necessary, and to ascertain in detail the market requirements for such products. It was then necessary to assess whether particular products were suitable for manufacture in the organization, or whether it would be better to buy them from outside. When a decision had been taken to manufacture a product it had to go through the full development stages, which included design, initial testing, and field trials. With the successful completion of the trials the production drawings would be issued and the equipment manufactured. The matter is then in the hands of the Sales Department to obtain orders for the equipment, which they had estimated to be in demand at the beginning of the development cycle. To illustrate this point, slides indicating the development of a range of multi-operator welding sets were shown.

The importance of the sales representative was stressed, for in many smaller organizations he is relied upon to give full technical information on the application of the welding process, and often is called upon to assist in the design of a component suitable for welded fabrication. It followed that for a technical representative to be successful he must have the full confidence of his customer and the ability to deal with his problems ably and expeditiously, from both a practical and technical viewpoint. With the introduction of more and more complex processes there was a continually increasing demand on technical representatives for a wider and deeper theoretical knowledge, in addition to sound practical experience.

One important feature of selling new welding processes was to introduce them to consulting engineers responsible for the design and specification of some of the major engineering projects, such as bridges.

In summing up his remarks, Mr. Boughton stated that successful salesmanship depended on giving the customer the service he wanted in every aspect, including technical ability of the product to meet his requirements, economics of the process being available, and the satisfactory delivery of spares and good service. There is no short cut to successful selling—the only way is hard work with a good product, which must be assisted by full Works' support, including advertising, exhibitions, and direct mailing campaigns. H.R.C.

Manchester — Gas Cylinders

Owing to the unavoidable absence of the lecturer, Mr. T. Reeks, the paper on "Safety in Welding" was read to members on 4th March by Mr. G. Williams and Mr. D. C. Bawden, two members of his staff.

Mr. Williams described the manufacture of cylinders for compressed gases, gave details of the materials used, and outlined the very strict regulations governing the manufacture and use of these containers. He pointed out that every cylinder for dissolved acetylene has to be examined once a year and all other cylinders every four years.

Complete records of every acetylene cylinder made are kept from the time it is completed until it is scrapped. Containers for different gases can be recognized by clear identifying colours: black for oxygen; maroon for acetylene; red for other fuel gases; grey for air, and grey with a black neck for nitrogen. Left-hand threads are fitted to all fuel gas cylinders; right-hand threads for other gases.

Mr. Williams emphasized the importance of correct storage of cylinders—oil and grease should be avoided, and acetylene cylinders should be kept in an upright position to avoid loss of the solvent acetone.

Mr. Bawden began his part of the lecture by describing the principles of high- and low-pressure blowpipes; he stressed the importance of adhering to the nozzle pressures recommended by the manufacturer. Good quality hose, specially made for welding and cutting purposes, should always be used, and it should be regularly inspected and kept in good condition. When repairs are needed the damaged portion should be replaced and only the correct fittings should be used for rejoining. There is danger in using odd pieces of metal tube for this purpose, particularly copper tubing for acetylene lines. All brass used in welding and cutting equipment operated on gas has to conform to a specification in which the copper content does not exceed 70%.

Mr. Bawden then explained the principles of single and two-stage regulators, and dealt with the problems of backfires and flashbacks—their causes and the action to be taken to avoid them. He warned of the dangers of gas leakage, both of fuel and of oxygen, which readily produce hazardous conditions. It was particularly important, too, to take special care when operating on tanks and vessels that had contained inflammable liquids or gases and when working in confined spaces. Mr. Bawden illustrated these remarks with examples of incidents that he had personally investigated.

The absorbing lecture was concluded with a short practical demonstration of how

flashes could travel back into hoses and how internal hose fires could be started. Like most other equipment, gas welding and cutting apparatus is perfectly safe as long as it is correctly used and maintained. But, as for example in electric wiring, there are dangers if high standards are not maintained.

J.M.W.

N.E. Tees-side — Annual Dinner

The Branch held its Annual Dinner at the Corporation Hotel, Middlesbrough, on Friday 20th March, when nearly a hundred members and guests were present to support the Branch Chairman, Mr. L. W. Cooley, who presided. The toast of the Institute was proposed by the Branch President, Mr. Hargraves, and the response was made by Mr. E. Seymour-Semper, Vice-President of of the Institute. The Vice-Chairman of the Branch, Mr. P. Bellamy, proposed the toast of the Guests, and the Chairman's health was proposed by Mr. A. W. Bell of the Tyneside Branch Committee.

South London — Design for Welding

The popularity of practical information on design for welding was amply demonstrated at the meeting held on 12th February, when about 190 members and visitors listened to a paper on "Welding and the Draughtsman", presented by Mr. H. B. Merriman. At least another 30 people had, with reluctance, to be refused admission. The success of this paper led to arrangements for a repeat lecture on 24th March, combined again with the North London Branch.

Mr. Merriman began with some comparisons of stresses in basic weldments; a procedure for combinations of static and dynamic loading of welds; and some observations on the design of ferritic and austenitic weld joints.

The necessary close association between the designer, metallurgist, welding engineers, and fabricator was assumed to be well known, and little time was spent on this vital aspect. Most of the material presented by the lecturer was drawn from his experience in the design of pressure vessels, but the substance could readily be applied to the design of general welded fabrications. The basic theme was that of 'designing-out' objectional features in welds, or where possible, 'designing-in' resilience or flexibility to the structure to preserve the welds from shock loads.

E.B.G.T.

High-Temperature Brazing

The Branch joined the Institute of Metals at their offices in Belgrave Square on 5th March to hear Mr. A. Cibula present a paper on "High-Temperature Brazing".

The author indicated that the more commonly known filler metals tend to lose their strength very rapidly at elevated temperatures, and that it is necessary to consider the nickel, cobalt, or silver-palladium type of alloys, which give a greater strength at the higher temperatures. He also showed that whilst the brazing temperatures might be of the order of 1050°–1250°C., the service temperature could be expected to be less than 400° below the brazing temperatures.

Most high-temperature materials owe part of their successful operation to the presence of a stable oxide film. But this film also inhibits satisfactory brazing, for the oxide prevents the wetting of the surface by the filler metal. For these reasons it is necessary to consider methods of fluxing. Mr. Cibula pointed out that the four usual methods were liquid fluxes, self-fluxing alloys, gas fluxes or vacuum methods, and plating. He considered in detail the strong points and weaknesses of these various methods. For example, he illustrated the value of plating on such materials as Nimonic 90, which had an aluminium content of about 2·5%. The success of this method depends frequently on not only the thickness of the plated layer but also on the strength of the bond between the plating and the parent metal.

In discussing the successful brazing of a joint, the lecturer mentioned various ways that are used to improve the penetration of the filler metal into the gap. One of these is the addition of alloying elements to the filler metal so as to provide some constituents that were soluble in the parent metal. This in turn would promote spreading. Another method, which is somewhat crude, is to increase the brazing temperature. A refined technique is to add a constituent, such as lithium for the copper alloys, to the filler metal. This has a self-fluxing action and although it does not alloy with any of the parent materials it improves the spread of the filler metal.

The influence of various brazing methods on the strength of the brazed joint was of interest. The liquid fluxes in one experiment provided considerably stronger joints than the gaseous fluxes; yet vacuum brazed joints were no stronger than those made with gaseous fluxes.

Mr. Cibula concluded his talk by stating that if brazing was to be successful, the selection of parent metal and filler material, the time and temperature of the brazing cycle, must all be considered against the particular component being brazed. If these matters are carefully considered and integrated then successful and satisfactory joints can be made.

E.G.B.T.

South Western — Aluminium Welding

A lecture on the "Inert-Gas Metal-Arc (Self-Adjusting) Welding of Aluminium, with particular reference to shipbuilding, was given by Mr. E. M. Wilson at Bristol on 6th March.

The lecture was preceded by a most interesting film showing the aluminium welding of the *M.V. Bergensfjord* deckhouses.

Mr. Wilson illustrated his talk with slides and showed some typical examples of both large and small ships' structures welded in aluminium alloy, together with modern trends in self-adjusting arc welding equipment.

The use of welded aluminium in the U.K. for this purpose has steadily increased in the past six years, due largely to the practical application of the self-adjusting arc process in the shipyard. This has been influenced by close co-operation between the shipbuilders, the aluminium fabricators, the welding equipment manufacturers, and the appropriate research and development organizations.

Much had been learned from the fabrication by welding in various shipyards of experimental ships' panels, and this work had formed the basis for the design and fabrication of larger structures.

There followed an interesting discussion, in which were highlighted such points as the practical measures taken to reduce distortion, the type of self-adjusting arc welding equipment necessary for both very thin and very thick materials, and the quality control of welds in aluminium and its alloys made by this process.

G.F.T.

Wolverhampton — Iron Powder Electrodes

At the meeting held on 18th February, Mr. A. B. Fieldhouse presented a topical paper on "Some Facts about Iron Powder Electrodes".

The use of electrodes containing a high proportion of iron powder is by no means a recent innovation, and the lecturer cited instances as long ago as 1933, when this method of obtaining high deposition rates was considered. It is only in the past few years however that this type of electrode has become popular, and Mr. Fieldhouse suggested that this was because of rising labour and production costs, and because the more recent electrodes of this type had a much wider field of application and were suitable for use by semi-skilled labour.

Mr. Fieldhouse went on to describe the seven different groups of iron powder electrode, classified according to main covering ingredients and the powder content. He stated that the rutile-based electrodes were becoming increasingly popular because of their all-position characteristics. The ductility of the weld deposit from this type of electrode tended to be lower than that of other types, but a recent development in which the combined water content was reduced to a minimum enabled exceptional mechanical properties to be obtained.

A popular misconception regarding iron-powder electrodes was that they have inherently better striking and restriking characteristics than the predominantly rutile coverings. The author showed conclusively that this was not necessarily so. In fact, neither material imparts instantaneous restriking through the electrode covering, even at a level of 80%. Similarly, it was shown that the mere addition of iron powder to a covering does not necessarily confer touch welding properties.

A number of slides were shown to illustrate the effect of iron powder additions on burn-off rate, metal recovery rate, welding performance, and weld-metal ductility. In regard to economies it was evident that iron powder electrodes will show to advantage in applications allowing a high operation factor, and particularly where labour and overhead charges are high. In these circumstances, despite the higher initial cost of the electrode, already offset by the increased metal recovery, a final saving in total welding costs is obtained.

After Mr. Fieldhouse had presented this most interesting paper, Mr. Stokes opened what occasioned to be a lively discussion period. Questions were asked about the welding of the medium carbon steels, deslagging in deep grooves, the economies of the twin-arc process, low-temperature impact properties, and a number of other aspects. Finally, after a request had been made for $\frac{1}{16}$ in. electrodes in the iron powder range Mr. Fieldhouse stated that he had

yet to find an operator able to use this size for more than a few minutes.

A hearty vote of thanks to the speaker for an exceptionally well presented paper was proposed by Mr. Dance, and this was unanimously supported.

For those who wish to study Mr. Fieldhouse's paper in detail reference should be made to his article published in the February 1959 issue of *Welding and Metal Fabrication* (p. 61). B.K.B.

Distortion

At the March meeting, a paper on "Locked-Up Stresses, Distortion, and Welding Procedures" was presented by R. J. Fowler, no stranger to most Branch members.

He introduced the paper by describing the circumstances in which welding causes locked-up stresses and distortion. The interrelation of these evils and the degree of joint restraint was clearly shown. In general, a low restraint results in low locked-up stress with maximum distortion; conversely, a high degree of restraint results in high stresses and less distortion. Restraint can be applied either by the particular situation of the joint in relation to the surrounding fabrication, or by the plate itself. Whilst the former could be controlled by a suitable welding procedure, it was impossible to do much to alleviate the effect of the latter, except perhaps by the judicial use of preheat. Typical shrinkages to be expected across fillet welds are 0·01–0·02 in. and across butt welds 0·05–0·10 in. Longitudinal shrinkage is not affected by plate thickness and amounts to 0·03 in. per 100 in. for fillet welds and 0·1 in. per 100 in. for butt welds.

To overcome some of the troubles of distortion and locked-up stresses Mr. Fowler recommended a number of useful 'hints':

1. Keep welds to minimum sizes and ensure that welders do not exceed them
2. Use the largest practical size of electrode and the fewest number of runs
3. Balance the welding
4. Use a well considered procedure, based on the building up of a structure from smaller sub-assemblies
5. Use pre-setting, preloading, preheating, peening, and post-heating only with great care and forethought, and under strict supervision, the requirements being largely based on previous experiences.

The author concluded his paper by describing with the help of slides a number of interesting examples of fabricated structures where these principles had been successfully applied. They included the mammoth fatigue testing rig orginally designed for the Britannia aircraft frame; some large fabricated bridge girders; frames for diesel-electric locomotives and their bogies; and the wind tunnel for the R.A.E. at Bedford. In all these instances it was evident that a tremendous amount of planning and preparation had been done in the welding design office before these jobs reached the shop floor, and also that a very close liaison had been maintained during construction to ensure the required accuracy of the end product.

A lively discussion followed in which the pros and cons of stress-relief, both before and after welding, figured prominently. It was agreed that the peculiarities of each case had to be considered individually, and that it was difficult to provide any hard and fast rules. Mr. Fowler reported that he could confirm the findings of others in regard to the use of high-metal-recovery electrodes that the consequent high welding speeds helped to reduce distortion.

In proposing a vote of thanks to the lecturer, Mr. Thompson said that it was refreshing to find someone who carried out in practice what is written in the text books on this subject. He complimented Mr. Fowler on a well-presented and practical paper, which had been most valuable to all who had listened to it. B.K.B.

CONFERENCES AND COURSES

The Electric Arc in Welding

Another one-day Conference in the series "Joining of Metals" is to be held this year in the Department of Industrial Metallurgy of Birmingham University. This year's subject is the electric arc in welding.

The Conference is being held at the University on **25th June 1959.** The programme is as follows:

9.30–9.45 a.m.—Opening address by Professor E. C. Rollason (Head of the Dept.)

9.45–10.30 a.m.—"Survey of Arc Characteristics and Their Significance in Welding" by University "Joining of Metals" Team (Presented by D. R. Milner).

10.30–11.0 a.m.—Coffee.

11.0–11.45 a.m.—"Gas Absorption from Arc Atmospheres" by University Research Team (Presented by Dr. G. R. Salter).

11.45–12.30 p.m.—"Mechanism of Metal Transfer in the Welding Arc" by University Research Groups and Electrical Research Association (Presented by J. C. Needham).

12.30–2.00 p.m.—Luncheon.

2.00–2.45 p.m.—"Heat Transfer from Arcs" by University Research Team (Presented by J. B. Wilkinson).

2.45–3.30 p.m.—"Plasma Projectors and Constricted Arc Torches: Their Design, Characteristics, and Application" by A. R. Moss, A.R.D.E., Ministry of Supply.

3.30–4.30 p.m.—Discussion.

4.30 p.m.—Afternoon Tea—End of Conference.

Applications for enrolment forms should be addressed to the Secretary, Department of Industrial Metallurgy, University of Birmingham, Edgbaston, Birmingham 15, and should be marked "Welding Conference". An enrolment fee of One Guinea will be charged; this will include the cost of luncheon and refreshments.

As in previous years, arrangements are being made for the papers presented at the Conference to be published in a future issue of the *Journal*.

Post-graduate course in structures and materials

Cambridge University's Engineering Department will hold its 1959–60 post-graduate course in Theory of Structures and Strength of Materials from 6th October 1959 to June 1960.

The principal subject will be the latest advances made in the understanding of the behaviour of metallic structures under static, repeated, or fatigue loading, with particular emphasis on welding as a method of fabrication.

The Course is open to graduates with industrial experience, and to others with the necessary qualifications and experience to obtain full benefit from a post-graduate course of study.

Further details and application forms can be obtained from the Secretary, Cambridge University Engineering Laboratory, Trumpington Street, Cambridge. The closing date for completed forms is 31st May 1959.

Welding for Foremen, Supervisors and Inspectors

In addition to the normal courses for welding operators and engineers Quasi-Arc Ltd. have arranged a special course of two weeks' duration for foremen, supervisors and inspectors. It will be held at the Company's Welding School at Bilston from 20th to 31st July 1959.

About half the time spent on the course will be allotted to technical and theoretical work dealing with arc welding procedures, electrode classification, distortion, welding of cast iron, radiographic examination, and supervision of welders.

The practical side, completing the course, will cover the welding of various types of joint, the different techniques connected with recently developed electrodes, and the testing and examination of welds.

Those wishing to take part in the course should apply for enrolment to Quasi-Arc Ltd., Bilston, Staffs.

BRITISH WELDING RESEARCH ASSOCIATION

Avtomaticheskaya Svarka

Following the announcement in the February issue of the *Journal* (p. 81) of the translation of this Russian periodical by the British Welding Research Association, and the distribution of the first specimen copy translation, the contents of the January and February issues (1959) have been made available as follows:

January

Foreword. Soviet welding in the seven-year plan, 1959–1965.

Choice and design of equipment for electroslag welding, *by P. I. Sevbo, V. E. Paton, and M. G. Belfor.*

Coiled flat aluminium tube, *by R. I. Barbanel, I. G. Martinov, and B. F. Lebedev.*

Vibration of the electrode: Its effect on droplet transfer of metal in electro-slag welding, *by D. A. Dudko and I. N. Rublevskii.*

Effect of aluminium on the porosity of automatic welds, *by L. S. Sapiro.*

Effect of prolonged ageing of austenitic deposited metal on its tendency to intercrystalline corrosion, *by M. A. Galperin and V. V. Ardentov.*

Determination by a magnetic method of the alpha-phase content of austenitic weld metal, *by N. I. Kopersak.*

Choice of metal and electrodes for welds to be used at low temperatures, *by I. N. Gerasimenko.*

Automatic welding of IKh 18 N 9 T steel intended for short-period use at high temperatures, *by E. Yu Yuganson.*

Radiographic investigation of the phase composition of austenitic-ferritic weld metal, *by N. Ya Karasik*

Local spectrum analysis for investigating the chemical non-uniformity of resistance welds, *by N. V. Bobritskii and N. A. Grinberg.*

Electrical impulse hard facing of metal deposited from liquid, *by M. S. Baranov.*

Effect on ingot properties of electro-slag heating of the riser by a graphitized electrode, *by D. F. Tchernega, V. A. Molotkov, N. N. Kisel, and K. G. Trofimova.*

Combined automatic welding of spherical containers for gas, *by V. R. Ryabov.*

Pacific Ocean Conference on Welding, *by V. N. Sherstyuk.*

February

Scientific and Technical

Heat resistance of welded joints in the nickel-chromium alloy Kh 20 N 80 T 3 Yu (EI 437 B), *by B. I. Medovar, A. N. Safonnikov, and R. O. Lents.*

Intra-crystalline segregation of phosphorus in electro-slag welds in carbon steel, *by Yu A. Sterenbogen, L. A. Poznyak, and G. I. Parfessa.*

Dynamics of the self-regulating arc welding process, using an inertia-type current source, *by V. M. Shchitova.*

Influence of cooling rate on the structure of deposited metal 3 Kh 2 V 8, *by E. N. Morozovskaya and G. I. Parfessa.*

Microstructure of aluminium weld metal, *by D. M. Rabkin and V. F. Grabin.*

Automatic hard facing of a layer of chilled cast iron by electrodes of ductile cast iron strip, *by G. V. Ksendzyk.*

Arc welding of stainless steel to carbon steel, *by I. N. Gerasimenko.*

Welding Practice

Use of gas cutting machines with template copying mechanisms for making welds of complex outline, *by A. E. Asnis and P. F. Baida.*

Discovery of a method of repairing a cracked area by single-side welding, *by M. M. Kraichik and V. B. Shlyapin.*

Improvement of the procedure for welding and heat-treatment of carbon steel wire for rope, *by G. P. Sakhatskii and R. M. Shirokhovskii.*

Electro-slag welding of flanges of explosion-proof casings, *by M. B. Tsukerman.*

News Section

Electro-slag welding of cast iron, *by Yu A. Sterenbogen and Yu N. Zaitsev.*

Automatic welding by exposed-arc of stable chromium-nickel austenitic steels, *by B. I. Medovar, A. N. Safonnikov, and L. G. Puzrin.*

The A–547–R semi-automatic machine for welding thin metal in a shielding atmosphere of CO_2, *by D. A. Dudko and A. G. Potapevskii.*

The annual subscription for twelve monthly copies of the cover-to-cover translations of this periodical is £10 10*s.* 0*d.* or $30, and should be sent to the British Welding Research Association, 29 Park Crescent, London, W.1, England.

OTHER SOCIETIES

SOUTH AFRICAN INSTITUTE OF WELDING

Awards

The Council of the Institute has announced the following awards for 1958:

The **Hall Longmore Award** to Mr. L. A. Woodworth for the best contribution made to the advancement of the science and craft of welding in South Africa.

The **Harvey Shacklock Award** to Mr. R. D. Gemmell for the preparation and presentation of his paper "Welding Repairs to Locomotives".

Both Awards were presented at the Eleventh Annual General Meeting of the Institute on 18th February 1959.

NEWS FROM INDUSTRY

Scottish Industries Exhibition

The third Scottish Industries Exhibition, promoted by the Scottish Council (Development and Industry), is being held at the Kelvin Hall, Glasgow, on September 3rd–19th, 1959. It will be devoted only to goods of Scottish manufacture. A number of welding equipment manufacturers and users of welding for fabrication will be displaying their latest products.

British Oxygen Company Ltd.

The directors' report of the British Oxygen Company, given at the annual meeting in London on 25th March, shows that group turnover rose by 15·7% during the preceding 12 months, and that profits were up by nearly £1¼ m., at £7,224,500.

Mr. J. S. Hutchinson stated that in the Electric Welding Division Quasi-Arc showed improved sales and profits, although import restrictions and keener competition overseas reduced exports. The automatic welding section had a busy year, with higher turnover and profits. New installations included the longest welding boom in the world, for use in constructing an atomic energy power station. A self-propelled double fillet welding machine, complementary to the existing gantry welding machine, was introduced successfully into the shipbuilding industry.

Group capital expenditure, totalling £24m. for the three financial years 1958/61, includes provision for additional capacity for the new large-scale process use of industrial gases by the steel, chemical and other industries.

Demand for industrial user equipment and industrial gases was down, both in the home and export markets, although sales of nitrogen and argon continued to rise.

Tool material

A new sintered tool shank material, Mallory NO-CHAT, has been introduced into this country by Johnson, Matthey and Co. Ltd., who claim that it has a similar coefficient of thermal expansion to tungsten carbide, so that tool tip cracking through thermal stresses in brazing the tip onto the shank is avoided. Data sheets are available from Johnson, Matthey's head office at 73/78 Hatton Garden, London, E.C.1.

Gas-shielded arc welding

The Rubery Owen Group has formed a Rowen Arc Division to manufacture under licence in Britain West-Ing-Arc equipment for inert-gas welding of aluminium.

A special laboratory is to be set up at Coventry for the examination of problems presented by industrialists and to offer advice on welding difficulties.

Computer-controlled cutting

Two British Shipyards, Swan, Hunter, and Wigham Richardson Ltd., and Vickers-Armstrong (Shipbuilders) Ltd., will be the first to install computer-controlled profile cutting machines, developed by British Oxygen Gases Ltd. in conjunction with Ferranti Ltd. Both companies have ordered prototype machines for installation in their yards next year.

Besides providing a considerable degree of automation in shipbuilding, the system is designed to achieve other important savings in the cost of production.

C. S. Milne & Co. Ltd.

Another of their memorable parties was given by the Company at the Palmerston Restaurant, Bishopsgate, on 3rd April. Mr. C. S. Milne presided over a gathering of some 250 friends, customers, and competitors (as he described them).

Mr. Milne proposed the toast of his Guests and the response was made by Mr. John Strong, President of the Institute of Welding.

An entertainment arranged by Sidney Jerome and Alfred Swain followed the Dinner, and added to the enjoyment of the guests.

Resistance welding control

At a meeting of the Merseyside Section of the British Institution of Radio Engineers on 20th February, Mr. C. R. Bates presented a paper on High Speed Electronic Welding Control. He described the development of NEMA standard type 3B sequence controls for resistance welding machines, giving the circuits and operational details of three types to show the progress from relay operation to fully electronic sequencing. He also described how negative hold-time was able to compensate the mechanical lags in machine operation.

The use of delayed firing of the ignitron contactor controlling the transformer allowed continuous current to be supplied starting at about 85° with respect to the supply voltage. A recent universal model capable of over 480 operations per minute was described.

Tubular Construction

Construction in steel tube is not cheap by comparison with normal steelwork in which all the joints are right-angle connections, but it does sometimes happen that right-angled joints are either undesirable or

impossible. In the Coronation Arch over the Mall in London, or in a crane jib tubular construction is elegant, light in weight, and strong. The joints are not always easy to make, but many types have been designed by specialists in steel tube. One such expert, Mr. G. B. Godfrey has written a most informative nine-page paper with 24 illustrations, in the *Structural Engineer*, April 1959.

Mr. Godfrey points out that the whole concept of tubular structures is based on welding, and that bolts are used only for site connections. Much of the article describes the requirements of the revised BS.449 which is due to be published in the summer.

Some of the sections in the paper concern: tubes of different diameter or thickness, tube reducers, end-to-end connections, saddle connections, eccentric and crimped connections, stability of the tube wall, diaphragm joints, polyhedral joints, shoes for trusses, column bases, bolted flange connections, and bracing.

The material in this paper forms a valuable supplement to the BWRA booklet "Welded Structural Details" for which Mr. Godfrey contributed some guidance.

J.S.S.

Proof stress determination

The properties of steels at elevated temperatures are of increasing importance, and to provide some degree of uniformity in measurements and methods of testing the British Standards Institution is preparing a series of Standards dealing with this subject. The first part of BS.3082 – "Mechanical testing of steel at elevated temperatures", gives dimensions for test pieces for determining proof stress and specifies the operating limits of the heating apparatus and the measurement of temperature. The recommendations are based on work carried out by a technical committee of the ISO.

Copies of the Standard cost 3*s.* from the Institution at 2 Park Street, London W.1.

Aluminium fabrication

For some years the Heaton Chapel Works of the Fairey Aviation Co. has been using inert-gas metal-arc welding techniques for the fabrication of equipment for the United Kingdom Atomic Energy Authority. Both the consumable electrode and the tungsten arc processes have been employed and welds in thick aluminium (up to 5¾ in.) have been made to meet the stringent requirements of the Authority.

One of the items that has been manufactured is a 4½ ton vessel known as the S250, part of which is shown in the accompanying illustration.

Production of aluminium fabrications at this branch factory has now reached 80–100 tons a month.

STUB ENDS

▶ A1 Electric Welding Machines Ltd. of Inverness have secured what they believe to be the largest order for British-built machine tool equipment placed by Czechoslovakia since the war. It is part of a project to establish a factory for the manufacture of motor-vehicle wheel rims on semi-automatic production lines.

▶ The British Electrical and Allied Manufacturers Association has enlarged and reprinted its booklet *Abbreviatons of Associations, Institutions, Research and Trade Associations.* One surprising ommission is the name of the Institute of Welding, though the qualifications A.M.Inst.W., and M.Inst.W. are given. Single copies may be obtained from BEAMA at 36–38 Kingsway, London, W.C.2, price 1*s.* 6*d.* There are reduced rates for bulk orders.

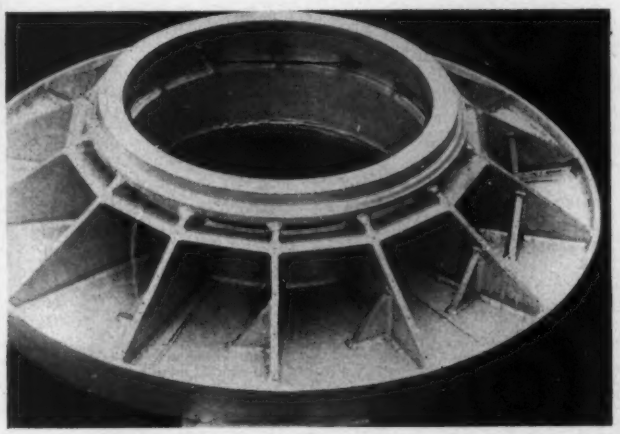
▶ The Metals Engineering Institute of the American Society for Metals has introduced a new comprehensive course on Arc Welding, designed for home study. Details can be obtained from the Institute at 7301 Euclid Avenue, Dept. N.R.–3, Cleveland 3, Ohio.

▶ Crompton Parkinson (Stud Welding) Ltd. have introduced a new range of rectifiers for use with their stud welding equipment. The larger units in the range are also suitable for arc welding.

▶ A universal oxygen hand cutter, the 'Hi-Lo' is being marketed by Centrajet Ltd. of York.

▶ Constructional work is proceeding on the new Camberley factory of Rockweld Ltd.

▶ Mr. C. A. Henderson, General Manager of Eutectic Welding Alloys Co. Ltd., has been appointed Managing Director of the Company.



An 8 ft 6 in. dia. welded fabrication in AW.5 aluminium alloy made by Fairy Aviation. The base is made from 1¼ in. thick rolled plate and the webs are about 1 in. thick.

▶ The April issue of *Light Metals* is devoted largely to the welding of aluminium. Contents include a guide to firms who weld aluminium, articles reviewing the work of B.W.R.A. and the British Standards Institution, and an article by leading metallurgists discussing the choice of different welding processes for various applications.

▶ The Central Electricity Generating Board is applying to the Minister of Power and the Local Planning Authority to build a nuclear power station at Oldbury-on-Severn, Gloucestershire, about 4½ miles downstream from the one now being built at Berkeley. The Board is also investigating three other sites on the Severn Estuary, at Tidenham, Aust and Portskewett.

▶ The treatment of metals by Bonderizing as a preparation for painting is described in a new brochure issued by the Metal Finishing Division of The Pyrene Company Ltd.

▶ Mr. A. C. Vivian, M.I.C.E., Managing Director of Metal and Pipeline Endurance Ltd., has recently joined the Board of General Descaling Company Ltd., an associated Company concerned mainly with descaling the insides of pipes and sewers both in the U.K. and abroad.

APPOINTMENTS

Situations Vacant

Metallurgist required by the Aluminium Development Association for work in the General Metallurgy Department. Duties include investigations into new applications of aluminium and its alloys, particularly with reference to welding and welded fabrication, answering technical enquiries, and preparing reports and reviews. Applicants should be Graduates in Metallurgy or hold the Institution of Metallurgists' qualifications, with some industrial or laboratory experience. Commencing salary will be according to qualifications and experience. Applications, giving full details, should be sent to the Technical Director, The Aluminium Development Association, 33 Grosvenor Street, London, W.1.

WM. DOXFORD & SONS (ENGINEERS) LTD. SUNDERLAND

FABRICATION SUPERINTENDENT

This appointment carries Assistant Works Manager status in Marine Diesel engine works and involves responsibility for the control of fabricating shops, modernization of methods and equipment, and liaison with the Technical Department on the design of fabricated structures. The range of work is wide and the post is an interesting one.

Essential qualifications are supervisory experience of fabricating shops in heavy industry, and a sound knowledge of modern methods of fabrication.

Applications, stating age, experience, and qualifications should be addressed to the Personnel Manager.

The cost of insertions in this column is 3*s.* 6*d.* a line,or 30*s.* per inch depth semi-display.

Box numbers are added for the additional charge of 2*s.* 6*d.* Replies should be addressed to Box. ——, Institute of Welding, 54 Princes Gate, London, S.W.7.

All other matters relating to advertisements for this section of the News should be addressed to the Executive Editor, 6 Ridge End, Hook Hill Lane, Woking, Surrey (Tel. Woking 2981).

Copy should be sent by 6th of each month for publication in the following month.

Current WELDING LITERATURE

Book Reviews
Additions to the Institute Library
Trade Publications

CONTENTS OF PERIODICALS RECEIVED

This section is intended to give a survey of the current welding literature received by the Institute of Welding Library. The contents lists are not exhaustive; only the main articles in welding journals are listed, and reprints from other journals and short notes are generally excluded. In addition, welding articles from other periodicals are listed. Abstracts of welding literature are given in the Bibliographical Bulletin of Welding and Allied Processes, *published by the International Institute of Welding, and details of this may be obtained from the Secretary of the Institute of Welding.*

Welding Journals

Canadian Welder, 1958, vol. 49, December

High-current automatic welding of aluminium (12–14)
Welding used extensively in all-aluminium refrigerator cars (18)

Canadian Welder, 1959, vol. 50, January

Continuous welded railroad rail (10–13)
Wall Colmonoy builds vertical brazing furnace (14)
Stainless steel used in missile design (16)
Fabricating a large stainless vessel (18–19)

Industry and Welding (U.S.A.), 1958, vol. 31, November

Production-line surfacing with tungsten carbide particles, R. Priest (32–33, 62)
Weld carbon steel to stainless with E316–15 electrodes (36–37)
Multiple arcs and powdered iron electrodes (38–39, 53)
Automatic oxy-acetylene welding makes quality tubing (40–43)
New Televised X-ray inspection (44)

Industry and Welding (U.S.A.), 1958, vol. 31, December

Welding procedures for fabricating aluminized piping (30–31, 58–59)
Automatic machine joins 168 wires to preformed tubing (34–35)
Welding aluminium without distortion (38–39, 41–42)
Eleven facts to know about ultrasonic soldering, R. Reynolds (45, 50)
Welding extends riveted stack (46, 51)
Aluminium-bronze overlay deposited without a defect, B. Ohmstede (48, 51)
Welding salvages a $38,000 die (52, 57)
Building beauty with welding (55–56)

Industry and Welding (U.S.A.), 1959, vol. 32, January

Semi-automatic process speeds welding of long deep-groove joints, L. J. McDonough (48, 56)
Automatic open arc welding at 150 ipm (51–52)
High speed stud welding on bridge beams (26–27)
Automatic equipment brazes and tests seven joints in four minutes (28–29)
A chart for arc welding twenty-seven types of stainless steel with eighteen different coated electrodes, W. E. McFee (30–31, 42)
Dual head automatic welds pipe assemblies (32–33)
Welded rockets are inspected with magnetic particles (34, 42)
Use both Gamma and X-ray equipment (36)
Welding Hastelloy alloy X to Inconel and to mild steel, C. Johnson (40–42)

Lastechniek (The Netherlands), 1959 vol. 25, January

Design against excessive plastic deformation, W. P. Kerkhof (1–9)
Recent articles in the field of welding (9–15)
Influence of internal stress on the behaviour of steel structures, W. Soete (15–19)

Oerlikon Schweissmitteilungen (Switzerland), 1959, vol. 17, January

New knowledge and development in metallurgy of welding in steelworks, K. L. Zeyen (5–131)

Le Pratique du Soudage (Belgium), 1959, vol. 13, February

A brazing and braze-welding process under gaseous flux, J. Delforge (25–29)
A few uses of welding in railway and tramway lines, Vandenberghen (31–34)

Przeglad Spawalnictwa (Poland), 1959, vol. 11, January

Problem of technical gas production in Poland, R. Sznerr (4–8)
Arc welding of cast iron, J. Wegrzyn (9–14)
Welding in steam turbine construction, R. Pasierb (14–18)

Current WELDING LITERATURE

Book Reviews
Additions to the Institute Library
Trade Publications

CONTENTS OF PERIODICALS RECEIVED

This section is intended to give a survey of the current welding literature received by the Institute of Welding Library. The contents lists are not exhaustive; only the main articles in welding journals are listed, and reprints from other journals and short notes are generally excluded. In addition, welding articles from other periodicals are listed. Abstracts of welding literature are given in the Bibliographical Bulletin of Welding and Allied Processes, *published by the International Institute of Welding, and details of this may be obtained from the Secretary of the Institute of Welding.*

Welding Journals

Canadian Welder, 1958, vol. 49, December

High-current automatic welding of aluminium (12–14)
Welding used extensively in all-aluminium refrigerator cars (18)

Canadian Welder, 1959, vol. 50, January

Continuous welded railroad rail (10–13)
Wall Colmonoy builds vertical brazing furnace (14)
Stainless steel used in missile design (16)
Fabricating a large stainless vessel (18–19)

Industry and Welding (U.S.A.), 1958, vol. 31, November

Production-line surfacing with tungsten carbide particles, R. Priest (32–33, 62)
Weld carbon steel to stainless with E316–15 electrodes (36–37)
Multiple arcs and powdered iron electrodes (38–39, 53)
Automatic oxy-acetylene welding makes quality tubing (40–43)
New Televised X-ray inspection (44)

Industry and Welding (U.S.A.), 1958, vol. 31, December

Welding procedures for fabricating aluminized piping (30–31, 58–59)
Automatic machine joins 168 wires to preformed tubing (34–35)
Welding aluminium without distortion (38–39, 41–42)
Eleven facts to know about ultrasonic soldering, R. Reynolds (45, 50)
Welding extends riveted stack (46, 51)
Aluminium-bronze overlay deposited without a defect, B. Ohmstede (48, 51)
Welding salvages a $38,000 die (52, 57)
Building beauty with welding (55–56)

Industry and Welding (U.S.A.), 1959, vol. 32, January

Semi-automatic process speeds welding of long deep-groove joints, L. J. McDonough (48, 56)
Automatic open arc welding at 150 ipm (51–52)
High speed stud welding on bridge beams (26–27)
Automatic equipment brazes and tests seven joints in four minutes (28–29)
A chart for arc welding twenty-seven types of stainless steel with eighteen different coated electrodes, W. E. McFee (30–31, 42)
Dual head automatic welds pipe assemblies (32–33)
Welded rockets are inspected with magnetic particles (34, 42)
Use both Gamma and X-ray equipment (36)
Welding Hastelloy alloy X to Inconel and to mild steel, C. Johnson (40–42)

Lastechniek (The Netherlands), 1959 ,vol. 25, January

Design against excessive plastic deformation, W. P. Kerkhof (1–9)
Recent articles in the field of welding (9–15)
Influence of internal stress on the behaviour of steel structures, W. Soete (15–19)

Oerlikon Schweissmitteilungen (Switzerland), 1959, vol. 17, January

New knowledge and development in metallurgy of welding in steelworks, K. L. Zeyen (5–131)

Le Pratique du Soudage (Belgium), 1959, vol. 13, February

A brazing and braze-welding process under gaseous flux, J. Delforge (25–29)
A few uses of welding in railway and tramway lines, Vandenberghen (31–34)

Przeglad Spawalnictwa (Poland), 1959, vol. 11, January

Problem of technical gas production in Poland, R. Sznerr (4–8)
Arc welding of cast iron, J. Wegrzyn (9–14)
Welding in steam turbine construction, R. Pasierb (14–18)

Some remarks on welding position determination, R. Duda (18–19)

Rationalized inventiveness facilitates welders' work, T. Ziemianski (19–20)

Modern oxygen cutting machinery in ship construction, R. Bechtle (20–24)

Souder. Revue pratique du Soudage (France), 1958, No. 19, December

Arc welding under flux: Lincoln machines (5–11)

Physiological effects of the electric arc (13–15)

The influence of gas on weld quality (17–20)

Arc welding with a consumable electrode in an argon atmosphere (21–31)

Schweissen und Schneiden (Germany), 1959, vol. 11, February

The behaviour of welded seams in the temperature range from −200°C. to 500°C., H. Hummitzsch (39–46)

New developments in the construction of resistance welding transformers, O. Gengenbach (47–51)

Welding in the peaceful uses of atomic energy, J. Ruge (52–60)

Schweisstechnik (Germany), 1959, vol. 9, January

Development of sintered flux showing an alloying effect, P. Beyersdorfer (3–4)

Experiences contributed by VEB Warnowwerft at Waremünde to extend applications of the Unionmelt welding method, U. Ploetz and W. Günther (5–7)

Carbon arc pneumatic planing, P. Nicht (8–10)

Possibilities of testing metallized layers, E. Kretzchmar and T. Drazkiewicz (11–13)

Acetylene generation techniques used in the Soviet Union, Shashkov (20–22)

Schweisstechnik (Germany), 1959, vol. 9, February

Factors affecting the design of welded constructions, especially of machine components under repeated load, J. Krebs (35–42)

Fatigue strength of CO_2 shielded butt-welds, A. Neumann and G. Müller (43–44)

Examples of the development of welded steel construction in crane manufacture, M. Koch (44–47)

The application of shell and box type construction in crane manufacture, H. Neumann (47–49)

Evaluation of X-ray films of welded joints by the five-notes-system, F. Wulff (49–53)

Possibilities of testing metal-sprayed coatings (recommended steps to standardize testing methods), part 2, E. Kretzschmar and T. Drazkiewicz (53–57)

Resistance and arc welding devices, H. Lück (58–60)

Saving of acetylene in the shipbuilding industry, G. Ratajczak (60–62)

First application of shielded CO_2 shielded arc welding in the shipbuilding industry, H. Fuchss (62–64)

The Welder, 1958, vol. 27, October–December

Nuclear power for the propulsion of commercial shipping, Sir John Cockroft (74–81)

The relining of concrete mixers by hard facing, J. W. Shedden (82–84)

Welding repair to the main engine crankshaft of a cargo vessel (85–87)

A guide to underwater welding, V. R. Foster (88–92)

Welding Engineer (U.S.A.), 1959, vol. 44, February

Precision welds in giant radio telescopes (29–31)

First all welded multi-storey building: a change in plans cuts costs of new Denver building, M. F. Yale (32–33)

A guide to nuclear power, part 3: metallurgical problems, W. D. Biggs (34–36)

Submerged arc builds California's first railroad bridge (37)

New storage vessel features thinner plate, improved capacity, R. V. McGrath (38–39)

Flame straightening of steel (40–43)

New pipeline welding method makes U.S. debut, T. Jefferson (44, 46)

Plasma heat–newest addition to welding's product lineup (50, 52)

Electronic tracing: an additional cost saver for production flame-cutting, L. S. McPhee (56, 58)

Welding Journal (U.S.A.), 1959, vol. 38, February

Fabrication of welded highway girders, R. A. Wilson (111–117)

Sterling-plus-lithium alloy selected for brazing B-58 stainless-steel sandwich panels, A. A. Lanzara and A. M. Setapen (118–124)

Automatic welding techniques speed erection of reactor vessels, W. A. Davis (125–128)

Training of inert-gas welders at an atomic power plant, G. J. Green and D. E. Callahan (129–131)

Electro-slag method of welding metals of large thickness, M. M. Barash, W. B. Heginbotham and P. B. L. Oxley (132–134)

Effect of backup variables on series spot welding of steel, E. F. Nippes, W. F. Savage and K. E. Dorschu (49s–56s)

Properties affecting suitability of 9% nickel steel for low-temperature service, T. N. Armstrong, J. H. Gross and R. E. Brien (57s–67s, and discussion, 65s–70s)

Development of titanium-clad steel using a vacuum-brazing process, D. Canonico and H. Schwartzbart (71s–77s)

Methods for end capping zirconium-clad reactor fuel pins, J. J. Vagi, R. L. Koppenhofer and D. C. Martin (78s–84s)

Brazing by the diffusion-controlled formation of a liquid intermediate phase, J. F. Lynch, L. Feinstein and R. A. Huggins (85s–89s)

Inspection of brazed joints: a statistical evaluation of some nondestructive methods, R. R. Ruppender and E. G. Baince (90s–96s)

Welding and Metal Fabrication, 1959, vol. 27, March

High speed production of light alloy tube, C. W. J. Vernon (80–82)

Modern refrigerator manufacture: a survey of metal-forming and welding methods at Kelvinator Limited, T. J. Palmer (83–91)

An assessment of gas-shielded I^2Rt welding, R. E. Jahn and L. M. Gourd (92–101)

Semi-automation in domestic iron production, D. D. Williams (102–103)

Control of welding distortion, E. H. Lee (104–108)

Welding Production (U.S.S.R.), 1959, No. 2.

Resistance welding of main pipelines, B. E. Paton and others (1–5)

Large diameter pipe longitudinal welds made by high-frequency resistance heating, N. P. Giukhanov and others (6–8)

Development of technology and equipment for welding non-rotatable joints of large diameter pipes, A. S. Falkevitch and others (8–12)

Experience in automatic welding of main pipelines, F. A. Zasko and G. E. Kogan (12–15)

The use of aluminium alloy pipes for gas pipelines, L. S. Livshits and A. S. Rakhmanov (16–17)

Improving the methods of control of gas and oil pipeline welding quality, I. E. Neufeld and A. S. Falkevitch (17–21)

The strength of gas pipeline welded pipes, M. P. Anutchkin and others (21–22)

Effect of fluxes on the properties of high strength steel welds, K. V. Lubavsky and others (23–25)

Automatic single electrode submerged arc surfacing of cylindrical products by a spiral method, F. F. Benoit (25–28)

Brass surfacing of ferrous metals by gas-flux technique, G. A. Asinovskaya and others (28–31)

Mechanical properties of welded and soldered joints of copper and copper-nickel alloys, V. R. Abramovitch (31–33)

Submerged arc welding of copper anodes under ceramic flux, K. V. Bagriansky and others (33–35)

The welding of very thick steels without bevelling the edges in hydrotechnical constructions, V. F. Dontchenko (35–37)

Experience with cold welding of cast iron, V. P. Riabinkin and S. V. Rulkov (37–42)

Raising the endurance of holders for underwater electric oxygen cutting, A. I. Schliamin (42–43)

Welding Research Abroad (U.S.A.), 1959, vol. 5, February

Researches on weldable high strength steels, H. Kihasa, H. Suguki and H. Tamura, Chapter 5: Weld ductility (2–39)

Some aspects of welding stainless clad steel cellulose digesters, G. Almqvist and S. E. Erikson (39–45)

I.I.W./I.I.S. Commission 3 "Resistance welding" Annual Report, Vienna Assembly, 1958 (50–53)

I.I.W./I.I.S. Commission 11 "Pressure vessels, boilers and pipelines" Annual Report, Vienna Assembly, 1958 (53–54)

I.I.W./I.I.S. Commission 13 "Fatigue testing." Tests to determine the influence of building-up with low-hydrogen electrons on the fatigue strength of medium carbon-steel shafts (55–59)

Zeitschrift für Schweisstechnik (Switzerland), 1959, vol. 49, February

The properties and weldability of new metals and alloys, C. G. Keel (32–36)

Various arc welding applications under shielding gases, A. Smarcan (37–40)

A study of MIG welding for light metals, H. Walser (45–47)

Zvaranie (Czechoslovakia), 1959, vol. 8, January

Welding techniques enter a new period, J. Cabelka (1–2)

Strength of some motor car parts under alternating stress, W. Gregor and O. Izdinsky (2–6)

Spotwelding of motorcar wheels for tubeless tyres, M. Lipa (6–9)

Equipment for tubular electrode production, K. Rejchlik (9–12)

New filled electrodes for surfacing the rolls of rolling mills, L. Lakatos (12–16)

Upsetting machines with electric resistance heating and examples of their application in mechanical engineering, T. Zrna (16–23)

Inspection of surfaced punched die edges, O. Zakavec (24–25)

Other Journals

Influence of the Peltier effect in resistance welding (*Philips Technical Review*, 1958–59, vol. 20, February, pp. 188–192)

Simple spray-brazing process makes tough jobs easy, H. E. Miller (*The Iron Age* (U.S.A.), 1959, vol. 183, 29 January, pp. 85–87)

Fabrication of refractory metals (conclusion), W. L. Bruckart (*Metal Industry*, 1959, vol. 94, 13 February, pp. 123–125, 132)

High temperature metal spraying (*Metal Industry*, 1959, vol. 94, 13 February, p. 130)

Palladium-containing alloys for high-temperature brazing, by E. R. Perry (*Engineering Materials and Design*, 1959, vol. 2, No. 3, pp. 134–136)

Plastic coated steel (*Engineering*, 1959, vol. 183, 13 February, p. 210)

Work measurement and productivity comparison in heavy arc welded construction, A. G. Thompson (*Engineering*, 1959, vol. 183, 20 February, pp. 242–245)

Aluminium and its alloys in 1958: some aspects of research and technical progress reported, E. Elliott (*Metallurgia*, 1959, vol. 59, February, pp. 79–85)

Post-weld annealing of stainless steel, M. W. Wilson (*Pipes and Pipelines*, 1959, vol. 4, February, p. 45)

Recent developments in inert-gas welding in the aircraft industry—2, R. C. Hesketh-Jones and R. L. Fannon (*New Zealand Engineering*, 1959, vol. 14, January, pp. 27–29)

An introduction to ultrasonic testing, by W. E. Schall (*The Inspection Engineer*, 1959, vol. 23, January–February, pp. 1–9)

Non-destructive testing as applied to the aircraft industry, W. G. King (*The Inspection Engineer*, 1958, vol. 22, November-December, pp. 129–133)

Welding and forming: development of special equipment for aircraft alloys (*Aircraft Production*, 1959, vol. 21, March, pp. 98–105)

Dielectric press "welds" vinyl-covered sandwich panels (*The Iron Age*, (U.S.A.), 1959, vol. 183, 5 February, pp. 83–85)

Advantages of hard-chromium plating in comparison with metal spraying and welding, W. Schmitt (*Werkstatt und Betrieb* (Germany), 1959, vol. 92, February, pp. 77–80)

Apparatus for testing the solderability of wire, J. A. ten Duis (*Engineers' Digest*, 1959, vol. 20, pp. 77–78, 82—from *Philip's Technical Review*, 1958/9, vol. 20, No. 6, pp. 158–161)

A new method of testing weld stresses, W. Rädeker (*Engineer's Digest*, 1959, vol. 20, February, pp. 83–86—from *Schweissen und Schneiden*, 1958, vol. 10, September, pp. 351–358)

Remote sealing of evacuated capsules, J. P. Evans and M. B. Waldron (*Nuclear Engineering*, 1959, vol. 4, March, p. 143)

Limit design of framed structures: characteristics of imperfectly plastic materials, J. Heyman (*Engineering*, 1959, vol. 187, 27 February, pp. 275–6)

ADDITIONS TO THE LIBRARY

BOOKS AND PAMPHLETS

ASA B 16.9-1958. *Steel butt-welding fittings.* New York, American Society of Mechanical Engineers. (Price $150)

ASSOCIATED OFFICES TECHNICAL COMMITTEE. *Rules for the construction, testing and cantlings of metal arc welded boilers and other pressure vessels.* 2nd ed. Manchester, A.O.T.C., 1958.

BRITISH STANDARDS INSTITUTION. BS.3067:1959. *Dimensions of blanks for seam welding wheels.* (Price 3*s*)
This standard has been produced to reduce the number of sizes of seam welding wheels. Only the outside diameter and the finished width of the blank have been specified.
BS.3065:1959. *The ratings of resistance welding equipment.* (Price 4*s* 6*d*)
The methods of rating described are for single-phase equipment operating under specified conditions of ambient pressure and temperature.
BS.1724:1959. *Bronze welding by gas.* (Price 6*s*)
The specific requirements of bronze welding are dealt with separately in accordance with methods for copper, mild steel, galvanized mild steel, cast iron, malleable iron, and for dissimilar metals joining. Appendices show correct flame, the removal of flux residues, and the chemical composition of copper alloy filler rods and wire.
BS.153:Parts 1 & 2:1958. *Steel girder bridges. Part 1—Materials, workmanship, protection against atmospheric corrosion. Part 2—Weighing and shipping erection.* (Price 6*s*)

CANADIAN STANDARDS ASSOCIATION. CSA B51-1957. *Canadian regulations for the construction and inspection of boilers and pressure vessels.* 5th ed. (Price 4*s* 10*d*)
CSA W59-1946. *Standard specification for welding of bridges, buildings and machinery (metallic electric-arc process).* 2nd ed. (Price 4*s* 10*d*)
CSA B63-1949. *Specification for welded and seamless steel pipe.* 2nd ed. (Price 7*s* 4*d*)
CSA W55.1-1950. *Specification for resistance welding equipment.* (Price 7*s* 9*d*)
CSA G30 series-1954. *Specifications for reinforcing materials for concrete.* (Price 9*s* 9*d*)
W117-1952. *Code for safety in electric and gas welding and cutting operations.* (Price 12*s* 2*d*)
CSA W48.2-1953. *Specification for corrosion-resisting chromium and chromium-nickel steel welding electrodes.* (Price 14*s* 7*d*)
CSA W55.2-1957. *Specification for resistance welding practice.* (Price £2 8*s* 9*d*)

"VODEX"

FROM FUNNELS...

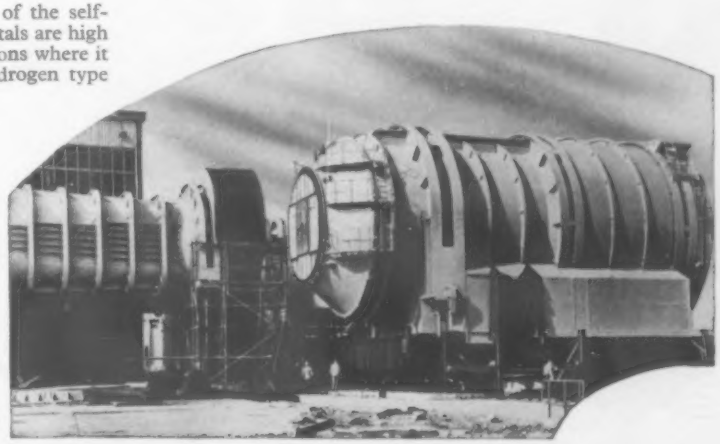
...TO TUNNELS

From funnels to tunnels—whether in a shipyard or on a construction job, or in the workshop, the Murex "Vodex" electrode is acknowledged to be the finest all-position electrode available. No other electrode of this type has ever succeeded in matching its all round properties and consistent quality. Indeed, many engineers will tell you that although they may have found one or two electrodes as good as "Vodex" in some respects, no other electrode can equal its versatile performance and fine quality.

The electrode has been designed for horizontal, vertical and overhead welding of mild steel and it is particularly useful where it is an advantage to have available one type of electrode capable of producing quality welds in all positions. While the composition of the flux enables vertical, upwards, and overhead welding to be carried out with ease, the "Vodex" electrode maintains all the essential characteristics for good downhand welding.

There is practically no spatter and the slag is of the self-inflated type. Physical properties of the weld metals are high and this enables "Vodex" to be used in applications where it might otherwise be necessary to use a low hydrogen type of electrode.

"Vodex" electrodes have been used for many world famous projects both at home and overseas. For example, "Vodex" electrodes have assisted in the construction of many famous ships, including the Royal Yacht "Britannia" built to Admiralty design by John Brown & Co. (Clydebank) Ltd. and the "Empress of England" illustrated above and built by Vickers-Armstrongs (Shipbuilders) Ltd., Walker-on-Tyne, for Canadian Pacific Steamships Ltd. "Vodex" electrodes were also used for all positional welding in the great wind tunnels built by Whessoe Ltd., Darlington, at the Royal Aircraft Establishment, Bedford. The techniques developed in the construction of these huge tunnels were later applied by Whessoe Ltd., to the building of the reactor pressure vessels at Calder Hall nuclear power station and again Murex "Vodex" electrodes were exclusively used. In short, from funnels to tunnels there is no finer all-position electrode available than "Vodex".



APPROVALS

Approved for use on mild steel in all positions by:

The Admiralty
Lloyd's Register of Shipping
American Bureau of Shipping
Det Norske Veritas
Complies with BS.639/1952 Parts 1 and 2
Accepted by Ministry of Transport
Accepted by Ministry of Supply, F.V.R.D.E. Specification 1050/1V
The electrode complies with the requirements of A.I.D.

"VODEX" THE FINEST ALL-POSITION WELDING ELECTRODE

MUREX WELDING PROCESSES LTD., Waltham Cross, Herts.

MUREX

T 31

TEXOLEX PROTECTS

The Texolex safety helmet and welding mask provides the welder with adequate protection against occupational risks. It comprises a high dome, cap type model No. 11 heavy-duty Texolex helmet, fitted with a swivelling and detachable one-piece fibre welding mask. It has two positive stop positions – down for welding, up for viewing – and is fitted with a shock resistant, non-arcing glass holder, a 4¼x3¼ inch, tinted lens to B.S. 679:1947, and clean cover glass.

Spare lenses and cover glass available, 4¼x2 inch tinted lens can be supplied.

MALCOLM CAMPBELL (PLASTICS) LIMITED

5 GREAT JAMES STREET LONDON W.C.1. Tel: HOLborn 5623 & 0931

RESISTANCE WELDING CURRENT METER

Designed to measure short pulses of heavy current

The standard instrument has 6 ranges covering readings of 500 to 100,000 amps. The readings are obtained by the use of toroids which can be split and placed around the heavy current conductor. Indication is stored enabling readings to be taken minutes after the weld is made. Can be used with welders employing synchronous and non-synchronous and energy storage controls.

HED

Illustrated leaflet sent on request

HIRST ELECTRONIC LTD.
GATWICK ROAD, CRAWLEY, SUSSEX

30 FT HEAVY STEEL ROLLS

CAPACITY AVAILABLE

Steel plates rolled any segment up to half cylinder, max. lengths, 30 ft in plate up to 1½ in. thick.

Cones and frustums developed and fabricated.

Formed plates supplied prepared for welding or completely welded.

Any type of fabrication undertaken steel or alloy, bar furnacing, bending, bevelling, electric welding, plate pickling, etc.

Marine structures up to 160 ft in length.

Keen prices, prompt deliveries.

W. A. PHILLIPS, ANDERSON & CO. LTD.
WEST HARBOUR ROAD, GRANTON
EDINBURGH 5

flexile flexuous flexible and tough

In other words — and in any length the job demands — Mersey Welding Cables are very strong and very supple, and therefore completely satisfactory for all welding operations.

They can be slung, hung or simply dragged — with the maximum of ease and manageability — over the roughest, toughest surfaces with no risk of damage to the hardy, resilient rubber sheathing and with the greatest freedom of movement for the welder himself.

With conductors of copper — or aluminium for extra lightness — and insulation of Vulcanised Rubber or P.V.C., they are cables of the highest quality produced at most competitive prices, backed by very considerable know-how and many years' experience of this particular industry.

Mersey Cables

All popular types of 'flexibles' are held in stock for immediate delivery; for more about these—and 'distributors' too — write for Leaflet No. 1/103

MERSEY CABLE WORKS LIMITED LIVERPOOL 20

A TI COMPANY

MCW 23

ARMOURED HEATERS

for the pre-heating and stress-relieving of welds and other forms of heat treatment.

Flexible resistance heaters to operate at working temperatures up to 800°C (1472°F) readily provide heat treatment to code requirements and have many advantages over other methods:

★ LOW INITIAL COST ★ LOW POWER CONSUMPTION
★ EASILY INSTALLED ★ EASILY CONTROLLED
★ SAFE LOW VOLTAGE ★ PORTABLE

Illustration shows typical set-up being prepared for heat treatment and welding
(photo by courtesy of Foster Wheeler Ltd.)

Please write for the new illustrated leaflet No. W/321

Electrothermal

ELECTROTHERMAL ENGINEERING LTD.
270 Neville Road, London, E.7 Tel. GRA 9911

SCHOOL OF WELDING TECHNOLOGY

courses may - october

D.18	Metal Spraying	4–6 May
D.7/1	Ultrasonic Inspection	25–28 May
D.15/1	Brazing Technology and Design	8–12 June
D.3/1	Welded Design and Construction in General and Mechanical Engineering	22–26 June
D.17	Advanced Course for Welding Engineers	22–30 September
D.1/2	Welded Design and Construction of Pressure Vessels and Pipework	12–16 October
D.12/1	Control of Distortion	26–28 October
D.13/1	Residual Stresses and Stress Relief	29–30 October

organized by

THE INSTITUTE OF WELDING, 54 PRINCES GATE, LONDON, S.W.7

KNIGHTSBRIDGE 8556

INVICTA *Overway*

ELECTRODES

British Code No. E.317. (BS.S. 1719)

THE INVICTA Overway electrode is designed to produce high quality welds when made in the overhead and vertical positions. It is also suitable for down-hand welding and has an excellent application in the making of fillet welds in rigid structures. It is specifically recommended for use on all mild and medium carbon steels, including steels to the following specifications: B.S.S.14, B.S.S.15, B.S.S.968, E.N.14.

Approvals

Complies with B.S.S. 639/1952, parts 1 and 2.

BRITISH

Admiralty.	All positions.
Lloyd's Register of Shipping (united with British Corporation Register).	All positions.
Ministry of Transport.	All positions.

NORWEGIAN

Det Norske Veritas—Mild Steel. All positions.

Also approved for welding "W" and "WW" Quality Steel.

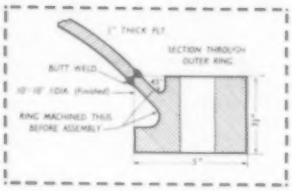
Used by some of the largest Shipbuilding and Engineering Works throughout the world.

Write for full details of the complete range of INVICTA ELECTRODES, which cover every industrial purpose.

Member of the Owen Organisation.

INVICTA ELECTRODES LTD. · BILSTON LANE · WILLENHALL · STAFFS · TELEPHONE: JAMES BRIDGE. 3131. EXTN. 303

The two mild steel heads for Dryers used in the paper-making industry are 12 ft in diameter and each weighs 2¾ tons.
The sketch shows how the solid ring has been machined out and prepared to receive the shell plates.
The segmental plates are pressings produced in our own works.



Welding by BOOTH

JOHN BOOTH & SONS (BOLTON) LTD., HULTON STEELWORKS, BOLTON
LONDON: 26 VICTORIA STREET, WESTMINSTER, S.W.1

Telephone BOLTON 1195
Telephone ABBey 7162

ADVERTISERS' INDEX

Anglo-Swedish Electric Welding Co. Ltd. . —
Booth, John, & Sons Ltd. 30
Braithwaite & Co. (Engineers) Ltd. . . —
British Industrial Gases Ltd. Inside front cover
British Insulated Callender's Cables, Ltd. 17
British Oxygen Gases Ltd. Outside back cover, 18
British Thomson-Houston Co. Ltd. . . —
Butterfield, W. P., Ltd. 12
Campbell, Malcolm (Plastics) Ltd. . . . 26
Carborundum Co. Ltd., The . . . —
Cooper & Turner Ltd. . . . . —
Courtburn Positioners Ltd. . . . —
Distillers Co. Ltd., The 2
Dorman Long (Steel) Ltd. . . . —
Electrothermal Engineering Ltd. . . . 28
English Electric Co. Ltd., The . . . 19
Eutectic Welding Alloys Co. Ltd. . . —
Flexello Castors & Wheels Ltd. . . . —
General Electric Co. Ltd., The . . . —
Gevaert Ltd. . . . . —
Hancock & Co. (Engineers) Ltd. . . . —
Harvey, G. A., & Co. (London) Ltd. . —
Head Wrightson Teesdale Ltd. . . 15
Hirst Electronic Ltd. 26
Ilford, Ltd. . . . . —
Imperial Chemical Industries, Ltd. . . —
Invicta Electrodes, Ltd. 29
Jenkins, Robert, & Co. Ltd. . . . 9
Johnson, Matthey & Co. Ltd. . . . 13
Johnson, Richard, & Nephew, Ltd. . . 14
Kodak Ltd. . . . . —
Lincoln Electric Co. Ltd. 22
Marconi Instruments Ltd. . . . . —
Marshall, Thomas, & Son, Ltd. . . 12
Marston Excelsior Ltd. . . . . —
Meritus (Barnet) Ltd. 16
Mersey Cable Works Ltd. . . . 27
Metal and Pipeline Endurance Ltd. . . —
Metropolitan-Vickers Electrical Co. Ltd. . 4, 5
Milne, C. S., & Co. Ltd. . . . . —
Morris, B. O., Ltd. . . . . —
Murex Welding Processes, Ltd. . . 25
Newton Victor Ltd. . . . . —
Norton Grinding Wheel Co. Ltd. . . 24
Philips Electrical, Ltd. . . Inside back cover
Phillips, W. A., Anderson & Co. Ltd. . 26
Quasi-Arc Ltd. 6, 7
Robey & Co. Ltd. 10
Rockweld, Ltd. 1
Saturn Industrial Gases Ltd. . . . 3
Siemans-Schuckert Ltd. . . . 11
Smith, Hugh, & Co. (Possil) Ltd. . . —
Spencer Wire Co. Ltd., The . . . 16
Stainless Steel Profile Cutters Ltd. . . —
Suffolk Iron Foundry Ltd. . . . 10
Welding Industries Ltd. 14
Wiggin, Henry, & Co. Ltd 20
Yates Plant Ltd. 8

Published by THE INSTITUTE OF WELDING, 54 Princes Gate, London, S.W.7
Made and Printed by Percy Lund, Humphries & Co. Ltd., London and Bradford

Solve your production problems with

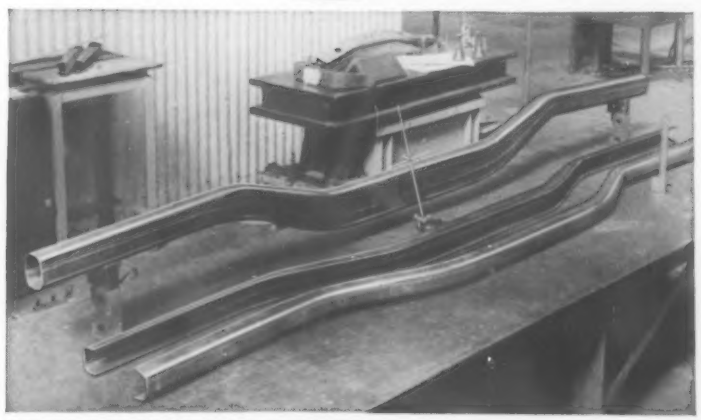
PHILIPS

bare wire

CO_2 welding machine

Philips AUTOMATIC CO_2 *bare wire* welding process is now available for immediate delivery in Britain. With this machine you can speed up mass production of welded parts, for example main chassis members, bottled gas cylinders, car wheels. And you can get welds protected from nitrogen and hydrogen by an atmosphere of CO_2 — which costs only a fraction of the price of argon. The absence of coating reduces the absorption of hydrogen, too.

- Welds very wide range of steel grades.
- Needs no stocks of expensive powders.
- Cuts labour costs by welding faster, requiring little attention in repeat operations.
- Welds are homogeneous with deep and even penetration.
- Arc is self-regulating . . .
- Nitrogen and hydrogen content low — no nitrogen porosity.



Component parts and finished vehicle chassis side member welded by Philips CO_2 process. *(Photo by courtesy of John Thompson Motor Pressings Ltd.)*

Philips Automatic CO_2 welding machine

provides automatic wire and CO_2 feeds. Travelling carriage moves along rails at speed set by operator. Water-cooled nozzle independent of main water supplies. Single control lever engages wire feed, switches on carriage motor and arc.

For further information about Philips CO_2 welding, and about Philips Automatic welding machine (a product of N.V. Philips, Eindhoven) write to:

PHILIPS ELECTRICAL LTD

INDUSTRIAL EQUIPMENT DIVISION

Century House, Shaftesbury Avenue, London WC2. GERrard 7777

(P10311)

Always ask for "ALDA" rods and fluxes

BRITISH OXYGEN SUPPLY ALDA

—the famous range of rods and fluxes. And a complete range of welding accessories— from goggles and gloves to friction lighters and wire brushes.

ALWAYS ASK FOR ALDA.

Write for fully illustrated literature.

BRITISH OXYGEN

British Oxygen Gases Ltd., Industrial Division, Spencer House, 27 St James's Place, London, S.W.1.

Outside back cover